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Life Prediction Methodologies for Composite Materials

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Life Prediction Methodologies for Composite Materials

Report of the Committee on
Life Prediction Methodologies for Composite Materials

National Materials Advisory Board
Commission on Engineering and Technical Systems
National Research Council

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NOTICE: The project that is the subject of this report was approved by the Governing Board of the National Research Council, whose members are drawn from the councils of the National Academy of Sciences, the National Academy of Engineering, and the Institute of Medicine. The members of the committee responsible for the report were chosen for their special competencies and with regard for appropriate balance.

This report has been reviewed by a group other than the authors according to procedures approved by a Report Review Committee consisting of members of the National Academy of Sciences, the National Academy of Engineering, and the Institute of Medicine.

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ABSTRACT

The committee assessed the methodologies used for the life prediction of organic composites utilized in load-bearing applications. The study is a state of the art survey of composites technology and life prediction methodologies for fiber-reinforced polymer composites under various loading conditions. Conclusions and recommendations are presented for future research and development to aid in the development of reliable life-prediction methods that allow the safe and predictable application of organic matrix composites in load-bearing structures.

PREFACE

As a result of continuing apprehensions in the United States about the life characteristics of composites and to realize the full benefits of composites use, the Department of Defense and the National Aeronautics and Space Administration (DOD/NASA) requested that the National Materials Advisory Board convene a committee to review current life prediction methodologies for composite materials and recommend the research directions with highest potential.

The committee found that despite the successful structural applications and demonstrations of composite structure in the fixed wing and rotorcraft industry, U.S. aerospace management still has strong misgivings about life characteristics of composite structure and the value of existing life prediction methodologies. The primary concerns are that a missed, or misinterpreted, failure process could possibly lead to catastrophic failure; the required predictive technologies do not exist; the predictive technologies, if they existed, would be too complex to permit designers to devise efficient structures within time and financial constraints; and the required material and structural characterization process would overwhelm the available resources.

A preliminary review of the status of composite life prediction methodology revealed a significant technology base that has not been assimilated by the development and procurement communities. Confronted by the conflicting views regarding both the fatigue sensitivity of composite materials and the availability of life prediction methodologies, the committee initially concentrated its efforts on gathering data to clarify the status of composites technology. Due to the breadth of the composites field and the interests of DOD/NASA, the committee believed that it could have the greatest effect by focusing on continuous-fiber-reinforced polymer-matrix materials for high-performance vehicles.

The committee defined life prediction methodology for composite materials as a set of procedures (or processes) and rules for attainment of an initial structural strength and retention of strength over a specified length of time. Thus, a life prediction methodology for composite materials is not a mathematical model (although it may include such models), but can be purely empirical, as is the case with some structural certification methodologies currently in use. The feature of our definition that is common to other methodologies is that it must be defensible by means of the scientific method.

While the committee emphasized that a sound lifetime prediction methodology should be based on an understanding of realistic failure mechanisms and modeling procedures that translate such understanding into practical design tools, it also recognized alternative procedures for achieving a life prediction methodology. Phenomenological methods of various types in use today, along with comprehensive experimental procedures, were reviewed and considered in the development of an outline of the type of model deemed desirable by the committee. The data were accumulated progressively and at increasing levels of material and structure complexity. The committee's findings are reported here, beginning with Background (Chapter 2), Constituents

(Chapter 3), Unidirectional Composites (Chapter 4), Structural Materials and Elements (Chapter 5), and ending with Structural Response and Design Practice (Chapter 6). Once the data were compiled, a collective committee opinion with respect to the status of life prediction methodology emerged. The results of the committee's deliberations are presented in Conclusions and Recommendations (Chapter 1).

The committee has not undertaken the task of defining the next composite material life prediction design paradigm. A significant data base has been collected and analyzed and some key issues and requirements for a new design paradigm have been established.

The committee would like to express its appreciation to Nozer D. Singpurwalla of George Washington University for his input on statistical issues in life prediction methodologies for composite materials. The committee also acknowledges with thanks the contributions of Robert M. Ehrenreich, Staff Officer and Pat Williams, Senior Secretary to the project.

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EXECUTIVE SUMMARY

There is an urgent national need for the generalized use of organic matrix composites in load-bearing applications. Organic matrix composites are generally stiffer and stronger on a unit weight basis than conventional construction materials and have the potential for better fatigue resistance, corrosion resistance, damage tolerance, design flexibility, and life-cycle cost reduction. Weight savings through the use of composite materials and the attendant improved vehicle efficiency and performance, is a technological advantage that is a determining factor in the international sales of aircraft and other high-performance vehicles. Since 1979, six foreign composite airframes and 10 major sets of airframe secondary components have been, or will shortly be, certified by the Federal Aviation Administration (FAA). During this same period, only three domestic composite airframes and 13 sets of airframe secondary components have been, or are in the process of being, certified by the FAA. Clearly, the United States faces serious competition in the development of advanced-technology commercial airframes and high-performance vehicles.

The review of applications for composite structures by the Committee on Life Prediction Methodologies for Composite Materials revealed that they can save weight and yield long life in both demonstration and production components. Composite applications to date have been designed to operate up to 4000 micro-in./in. strain in fixed wing and rotorcraft loading environments. In the fixed wing industry, these structures function at this strain level sufficiently beyond lifetime design goals to suggest that the need for life prediction methodologies is not a critical issue. However, the committee also observed that aerospace management still has strong misgivings about the life characteristics of composite structures and the availability of life prediction methodologies. The primary concerns are that a missed, or misinterpreted, failure process could possibly lead to catastrophic failure; the required predictive technologies do not exist; the predictive technologies, if they existed, would be too complex to permit designers to devise efficient structures within time and financial constraints; and the required material and structural characterization process would overwhelm the available resources.

Due to the existence of conflicting views regarding the fatigue sensitivity of composite materials and the availability of dependable life prediction methodologies, the committee initially concentrated its efforts on gathering a supportable data base to clarify the status of composites technology. The data base was accumulated progressively and at increasing levels of material and structural complexity (i.e., constituents, unidirectional composites, structural materials and elements, and structural response and design practice). Once the data base was compiled, the committee reached a consensus concerning the paradigm deemed desirable for further development of life prediction methodologies.

The paradigm centers on the recommendation that composite materials should be modeled at a level that reflects microstructural effects on strength and lifetime characteristics. The desired modeling concept emerging from the research community is that composite material damage state, damage growth, and strength can be determined by characterizing a small number of failure modes, each of which may be modeled within a representative volume unit that contains the failure process. These failure modes and related volumes are being defined at the research level but are yet to be developed to the point where they would be of use to the production designer. The potential to incorporate emerging tools (e.g., constituent materials characterization, failure state characterization, failure process scaling, statistical characterization methods, large-scale structural simulations, and simulated service environment characterization) leads to the determination of three findings where new or intensified technical efforts could result in significant gains, measurable in terms of both improved structural performance and increased user confidence.

FINDING No. 1: REDUCTION IN POTENTIAL FOR STRUCTURAL FAILURE

The potential for failure of composite structures could be reduced by the completion of a comprehensive analysis procedure that connects the understanding of basic local failure mechanisms with overall structural failure. The committee has the following two recommendations.

- Four areas of research where further efforts should be concentrated to produce the technology required are completion of analytical and experimental efforts to define the strength and lifetime characteristics for the axial fiber direction compression fatigue failure mode; completion of analytical and experimental efforts to understand failure mode coupling through structural collapse; development of simplified design and analysis procedures to permit failure mode characterization techniques and design rules to be introduced into vehicle design at the predesign level; and use of the failure mode characterization strategy to provide designers with a balanced view of emerging materials such as thermoplastic matrix composites. Completion of this research effort and formulation of design, analysis, and test procedures based on this work will reduce the magnitude of the total task and yield a greater degree of success at the application level.

- The representative volume element analysis, coupled with sufficient proof that the failure process list is exhaustive, must be properly packaged to instill greater confidence in producers and users, and cause increased use of composites for structures.

FINDING No. 2: RECOVERY OF LOST PERFORMANCE

The full potential of composites in high-performance vehicles could be attained with the development of new design rules, but current design practices for composites are based on traditional material forms (i.e., tape, fabric, and roving). Also, since specific knowledge of the lifetime characteristics of composite materials has progressed faster than design practice, great opportunities exist for product improvement. The committee developed the following recommendations.

- Opportunities for product improvement and cost reduction exist through broader consideration of materials forms (e.g., three-dimensional preforms, pultrusions, preplied cross-plied tape, and braiding). This broader view of composite materials must be included to influence the failure mode characterization research effort.

- The loss in performance of composite designs can be recovered by coupling the material with a structural design methodology such that the secondary failure modes are removed while maintaining the high-efficiency, fiber-dominated performance.

FINDING No. 3: REDUCTION IN RESOURCES NEEDED FOR CHARACTERIZATION

The formulation of analyses for composite structure based on failure mode characterization could reduce the scope of testing required to develop and certify composite structures and instill greater confidence in them. Number, size, and complexity of test specimens could be reduced by defining specimens to test each minimal representative volume element; using damage characterization for each specimen to maximize the information from each test; relating damage state to property changes within the test objectives (e.g., strength or stiffness); using statistical characterization at the representative volume unit level; and relating the spectral content of each test to the user spectra. The committee's recommendation is as follows:

- To reduce the effort required to develop a component, the scope of the life prediction process must be narrowed to specific failure modes and the development of techniques must concentrate on mapping between material systems.

Accomplishment of these recommendations should satisfy the conditions required for the expanded application of composite materials. Increased and more effective use of composite materials will produce, by means of improved vehicle performance, an improved capability for the United States to remain internationally competitive in the production of high-performance vehicles.

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CONCLUSIONS AND RECOMMENDATIONS

On the basis of the data accumulated for this report, the committee reached a consensus concerning the emerging paradigm deemed desirable for the further development of life prediction methodologies. The resulting methods are a reality in the sense that they are supported by research results. Applications engineers are aware of the new methods, but significant efforts are required to complete the models for each failure mode and to implement new design and analysis techniques in a format amenable to product design. An overview of the consensus model follows.

RECOMMENDED LIFE PREDICTION MODEL

While the number of new material systems and permutations of those materials in application-specific forms is bewildering, the resulting application-specific forms exhibit a limited number of critical local-strength and fatigue failure modes. These failure modes have been identified as fiber bundle axial failure, matrix microcracking, interlaminar disbonding, and local compressive instability. Each of these failure modes is associated with a representative volume of the composite that contains the failure region, and each occurs throughout the material in a dispersed fashion. Each mode has been found to exhibit a distinct subcritical growth mechanism and rate, and each also has an equally distinct instability process, both in terms of failure mode and structural criticality. The recommended concept for lifetime modeling is to evaluate both the accumulation of subcritical damage from each mode and the interaction of modes as dispersed throughout a structure; to determine the change in local properties resulting from the subcritical damage; and to assess the effect of the total distributed damage upon the functional integrity, stiffness, and strength of the structure. While stiffness will monotonically decrease as new surface area is created by means of subcritical damage, strength has been demonstrated to increase or decrease depending upon the material and structural configuration.

This model has been utilized in varying forms by many investigators over the years, and substantial progress has been made in dealing with limited laminate classes and loading conditions. The interaction of benign modes (e.g., transverse cracking) and potentially dangerous modes (e.g., fiber tension, compression, and delamination) is less well developed in the sense of impact upon structural, functional integrity.

Much remains to be done to extend this type of modeling to the case of general material environmental loadings and possible critical failure mechanisms. Furthermore, even successful completion of the general models would result in a deterministic representation of life for a process that is known to be stochastic with large variances. It should also be noted that the attention that has been given to various failure mechanisms, laminates, and environmental loads

conditions has largely been driven by existing materials. The relative importance of various failure mechanisms may very well be different when new materials are developed. In fact, the specific knowledge derived from existing materials could well drive the development of new and highly fatigue-resistant materials, even with respect to benign failure modes. *Despite the limitations in the model that are largely due to the state of development, the committee forecasts significant benefits to be derived from changing the technical point of view to one that captures the impact of microstructure on the strength and lifetime characteristics of composite materials.*

TECHNICAL OPPORTUNITIES

Since new and specific knowledge of the lifetime characteristics of composite materials has progressed faster than design practice, new opportunities for product improvement and certification now exist. Three technical opportunities were identified where new or intensified research efforts could result in improved structural performance and increased user confidence.

Reduction in the Potential for Structural Failure

The committee agreed that the potential for failure of composite structures could be reduced by the completion of a comprehensive program that connects the understanding of basic local failure mechanisms with overall structural failure. Four areas of research have been identified for which further efforts should produce the technology required. They are: completion of analytical and experimental efforts to define the strength and lifetime characteristics for the axial fiber direction compression fatigue failure mode; completion of analytical and experimental efforts to understand failure mode coupling through structural collapse; expansion of the failure mode characterization efforts to provide a basis for understanding the effects of thermal and chemical environments; and use of the failure mode characterization strategy to provide designers with a balanced view of emerging materials, such as thermoplastic-matrix composites. Within each of the four areas of research, there are specific problem areas that require high priority.

Axial fiber direction compression fatigue is the least understood failure process in a composite. Mechanistic models have not been correlated for this mode. Since the logarithmic slope of repeated-load strength reduction data is low for this process and the ability to carry compressive load is critical in many aerospace applications, it is of the highest priority for future investigation.

Although most failure modes have been modeled, modeling of failure mode coupling through structural collapse has not been demonstrated for significant structures. This technology should be emphasized because it is one of the most critical steps toward gaining designer and management confidence in the technology. To date, failure mode growth and failure mode coupling remain complex research-oriented tools. As the tools are used, insight derived from examination of test problems should permit generation of simplified techniques (including design rules for simultaneous material and structure design) that are amenable to incorporation into preliminary design. Analysis techniques need to be further developed that, by failure process, relate coupon data to full scale structure. Further understanding of the effect of damage and damage scale on structural response should help remove many current correlation problems between coupons and structure, such as compression strain failure after impact of coupons being significantly lower than that for structural elements; fatigue life curves changing slope as coupon geometry changes; wide coupons failing at longer fatigue lives than narrow coupons because edge

induced delamination stops growing; and fatigue damage that leads to failure, starting in the center of large unnotched panels and not at the edge as in narrow coupons.

The environment, for purposes of life prediction for composite materials, includes those factors that impact service life to the first order. To different degrees for different composite materials, the environment includes mechanical, thermal, and chemical "loads." Significant efforts resulting in a sorting of relative weights for the composite material life capacity for components of the environment were found at the materials level (Chapter 2). The same sorting was also seen at the structural level (Chapter 5), in terms of both analysis and test efforts, and concerns. Because of the time taken to achieve and technically accept understanding of composite damage and damage growth, the development of the understanding of the coupling of environmental effects on composite life characteristics has lagged. It is recommended that research be conducted to begin to understand the chemical and thermal environmental effects on specific failure modes. This work must include appropriate physical damage characterization, coupled with environmental exposure, and should culminate in combined effects damage growth simulation and modeling. Environmental effects characterization must also include a deliberate and quantitative understanding of the effects of time on the failure processes, which is especially important for understanding thermoplastic resin and pressure vessel applications.

Despite the excellent static toughness of thermoplastic matrices, data indicate that the time-dependent properties (e.g., crack growth rates, tearing energy reduction with fatigue cycles, and creep) will result in poor lifetime characteristics. Further characterization and study of the impact of the time-dependent properties of thermoplastic matrices on lifetime characteristics of thermoplastic-matrix composites should be undertaken before aerospace applications are expanded.

Although the new modeling methods can be used to reduce the potential for structural failure and to increase confidence in the structures produced, the ultimate role of life prediction methodology is the inversion of the analysis: *to develop design rules to guide material and structural configuration selections that will have long failure-free operating lifetimes.*

Recovery of Lost Performance

Current design rules based on now traditional material forms are too restrictive for life optimization of composite designs. Today's designs preclude three-dimensional material forms, restrict laminate orientations, treat notched strength by "brute force" reduction in strain, reduce strain to compensate for transverse or interlaminar failure processes, and group plies to optimize producibility. These restrictions have led to applications with marginal structural efficiency payoff. It is the consensus of the committee that the lost performance is recoverable by coupling the material and structural design such that the secondary failure modes can be minimized while maintaining high-efficiency, fiber-dominated performance. The understanding of failure process needed to accomplish this goal is either already available or is rapidly emerging in the scientific literature. However, organizing the data in a form tractable to designers is yet to be accomplished.

The payoff from coupled materials and structure design is to raise structural efficiency and accomplish damage containment efficiently early in the design process. To this end, the committee recommends the development of simple analytical models that allow: prediction of whether damage will grow under static loads; understanding of damage containment mechanisms; and prediction of material forms and lamination sequences that give the most advantageous stress states.

It is recommended that a simultaneous material-structural design process, based on understanding the critical failure mode concept, be used in the design and testing of structural components. These efforts should be integrated from the materials through structural test functions. Demonstration structures will provide the basis for both the next-generation production hardware and design technique.

Accumulation and organization of the data required to design and certify a structure do not have to overwhelm the design process. *By means of narrowing the scope of the life prediction process to specific failure modes and concentrating development on techniques to map between material systems, the effort required to develop a component can be reduced.*

Reduction in Resources Needed for Characterization

Tailoring characterization and development programs using metal structure paradigms has resulted in resources being misapplied. Because of the limited number of distinct failure modes, concentration on characterization by failure mode could greatly reduce the scope of testing required to develop and certify composite structures. Five research efforts were identified by the committee that would form the basis for actualizing a reduction in resources required for characterization. The recommended efforts include definition of test specimens at the scale and complexity of minimum representative volume units; use of damage characterization methods to define specifically the physical damage process for each specimen; definition of the techniques required to relate damage state to property changes; characterization of the statistical characteristics of each process at the representative volume unit level; and definition of a program to resolve, for each failure mode, whether life prediction under user spectra can be defined using constant-amplitude data.

For the case of the most highly developed failure process, axial fiber tension, the scale of the failure process has been demonstrated to be very small. For failures where fracture cleaves fibers and the representative volume element is very small, strength has been shown to scale through failure of very complex specimens that contain stress concentration, even including mechanical fastening. Even in this highly developed case, the community, in general, does not use the results to reduce the scale and complexity of materials characterization. The committee recommends the conduct of research and experimental programs to define the relationship of failure process and associated representative volume element scale to define minimal characterization processes for the larger scale, and more complex failure modes, such as axial fiber direction compression delamination and transverse cracking.

Specific knowledge of the damage state in the characterization tests forms the basis for translation of the results to structures and formulation of a certification test. In the past, many programs conducted life characterization without precisely defining damage state. The committee recommends that the physical character of composite damage and damage growth be emphasized. State of the art damage detection and characterization methods need to be incorporated in all lifetime characterization programs. Life prediction is a poorly defined problem. Emphasis should be placed on characterization of physical damage state or property changes, such as strength and stiffness, all of which can be related to application performance requirements. There should be increased emphasis on the incorporation of statistical representation of damage state and damage growth into mechanistic models. The statistical character of a composite fatigue process is usually a unique signature of the process and will be required to extend mechanistic models to unfamiliar circumstances. Statistical correlation at the mechanistic level should provide a significant opportunity to improve certification techniques.

The capability to integrate successfully and analytically through complex load histories has been demonstrated only for the case of interlaminar crack growth. For interlaminar crack growth, power law matrix cracking can be characterized and analytically integrated using energy release rate methods. The issue still exists as to whether constant-amplitude data are appropriate for fatigue characterization in general. It has been demonstrated that the flaw growth rates in laminates are altered by stress history. It should be noted that, while constant-amplitude testing remains the load history of choice for laboratory work, structural characterization for certification currently is based on random-amplitude testing for all of the federal agencies surveyed by the committee. The committee recommends further emphasis of alternative environment characterizations for laboratory efforts (e.g., random-amplitude testing using spectra that can be easily scaled for spectrum changes by means of spectrum ratio-endurance relationships) to circumvent the much larger problem of development of constant-amplitude data bases.

As an adjunct to the development of characterization procedures that conserve resources, analysis procedures need to be developed and demonstrated that will correlate, by means of failure modes, the translation of coupon data to structures. *These demonstrations will further confirm that appropriate changes in laboratory procedure can produce efficient and exhaustive composite characterizations.*

COMPETITIVE STATURE OPPORTUNITY

The early production commitments for composite demonstration hardware were made on the basis of accelerating actualization of potential weight savings. Unknown problems had to be addressed during the execution of these important first demonstration steps. Resolution of these problems (i.e., sensitivity to impact damage and water absorption) was met by accepting lower usable strengths as driven by schedule and funding constraints.

Current production commitments are being made on the basis of weight savings at a demonstrated life capacity. Life capacity is determined by means of full-scale combined environmental and load effects testing. Although expensive, these tests are producing confidence in the use of composites in new applications. Broader applications to primary structure will require reductions in the scale and scope of testing and a greater reliance on analysis together with more aggressive use of the inherent strength and stiffness of composite materials.

Additional hardware application opportunities could be actualized by removing the primary concerns of the aerospace management. The research community is on the threshold of proof that composite failure modes are exhaustively characterized. The applications community has achieved a heuristic understanding of the most subtle of the transverse cracking failure processes and have begun to implement appropriate material design rules to reduce both flaw development and environmental sensitivities by logarithmic amounts. With the single exception of damage introduced by uniaxial compression loading, understanding of the failure processes is sufficiently complete to sustain a new era of enhanced efficiency and life for composite structure.

Ameliorization of subcritical matrix cracking should result in life being limited by the environmental stability of the constituent materials. In the case of epoxy resins, a decade and a half of environment-exposure testing by NASA has shown epoxy matrix composites to have excellent environmental resistance with no "hidden" long term failure processes.

Proof that the failure modes characterized are exhaustive should remove the possibility of a missed or misinterpreted failure process that could lead to catastrophic failure. Existence of the analytical capability to relate coupon failures to structural failure should prove the existence of the required predictive technologies. Simplification of the predictive technologies to design rules applicable to preliminary design, should permit design of structures within time and budget constraints. Implementation of a representative volume-unit-based characterization technology should prevent the available resources from being overwhelmed. *Accomplishment of the above-cited technology developments should satisfy the conditions required for expanded composite applications, and the increased utilization will produce, by means of improved vehicle performance, an improved capability for the United States to remain internationally competitive in the production of high performance vehicles.*

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BACKGROUND

The intensive development of organic matrix composite materials for high-performance structures has been under way for more than three decades. Throughout this period, the field of advanced composite technologies has viewed the understanding of failure processes in composites as an extremely important problem that requires further development. As early as 1967, a National Materials Advisory Board report entitled Structural Design with Fibrous Composites stated that government programs should experimentally pursue definitions of elastic, inelastic, fatigue, and failure mode behaviors for composite materials. It was also recommended that such experiments should "correlate the results of tests of components, elements, and material specimens and be closely coupled with a strong effort in theory and analysis" (National Materials Advisory Board, 1968, p. 6). Twenty-one years after these recommendations, however, another National Research Council committee concluded that the use of advanced organic composite materials in production aircraft structures has been extremely slow and that the lack of adequate life prediction methodologies is a key inhibitor to the expanded utilization of composite materials (Aeronautics and Space Engineering Board, 1987).

The corrosion resistance, weight savings, and attendant improved vehicle efficiency enabled by the use of composites are technological advantages that are determining factors in the international sales of aircraft and other high-performance vehicles, however. Since 1979, six foreign composites airframes and 10 major sets of airframe secondary components have been, or will shortly be, certified by the Federal Aviation Administration (FAA). During this same period, only three domestic composite airframes and 13 sets of airframe secondary components have been, or are in the process of being, certified by the FAA. Clearly, the United States now has serious competition in the development of advanced-technology commercial airframes and high-performance vehicles.

The committee identified four factors, either intrinsic to composites materials or resulting from technology implementation assumptions that are key to the delayed development and acceptance of composite life prediction methodology. This background information is important to understanding the point of view of this report. The four factors are: implementation of design and analysis paradigms that neglect the effects of microstructural detail on the macroscopic response of composite materials; perception of the need to characterize fully the bewildering number of material systems available and their permutation, by means of material orientation; lack of consensus concerning failure modes and failure criteria among government agencies and technologists; and persistent use of design and analysis paradigms based on metals technology.

IMPORTANCE OF COMPOSITE MICROSTRUCTURAL DETAIL

A currently accepted practice in the composite design and analysis community is to use the single ply as the engineering scale for assuming homogeneous behavior. This assumption stems from the fact that metals can generally be treated as homogeneous at the engineering scale without regard for intergranular characteristics. Unfortunately, this assumption is not verified by the observations of microstructure and microstructural-scale effects on the macroscopic fatigue response of composites.

Reversal of the homogeneous ply assumption was key to the development of mechanistic fatigue models. New mechanistic models based on observations and modeling of microstructural effects on flaw field propagation now permit simulation of fatigue processes in composites. In addition to enabling fatigue modeling, consideration of microstructural detail will prove to be important in the reoptimization of composite materials for strength and life.

NUMBER OF MATERIAL SYSTEMS AVAILABLE

Advances in both fiber and matrix properties have occurred rapidly. When the number of new systems are permuted with the application-specific forms available, the result is a bewildering number of partially characterized materials. The performance advances have been significant enough that specification of a "best material" has been attractive only for a specific application at a specific time. Characterizations for applications tend to emphasize problems concurrent with the application and are never "complete". Thus, it has become obvious that the reduction of life prediction technology to a tractable state will not occur by cataloging a few characteristics for a nearly infinite number of materials. There have been significant gains at the research level in understanding the character of defects and defect growth in composite structure. Definition of the scale of the structural element required to characterize defects and defect growth has provided the opportunity to develop methods to map between material systems based on the quantitative understanding of failure process, utilizing minimal specimens both in number and scale.

VARIATIONS IN FAILURE CRITERIA

Wide variations in failure criteria exist among the government agencies that procure and certify composite structures. Furthermore, there is little evidence that the understanding of defects, defect growth control, and defect-effects on useful life is available in a format that will guide material and structure design. Defect sizes associated with various failure criteria range from no defects visible to the naked eye, to the acceptability of large defects that can be proven not to grow. Technical linking of defect state to useful life is usually not accomplished. Part of the problem may be that life prediction itself may be a misnomer. Time to failure is often a poorly defined technical concept, especially since composites may exhibit benign but easily visible "failures." Useful life may be better defined as changes in observables, such as damage state, stiffness, or strength. Such property changes can be interpreted by performance simulation models that relate such changes to functional performance, reliability, and safety.

DESIGN PARADIGMS BASED ON METALS

Current material forms (i.e., tape, fabric, and roving) have matured in a process of natural selection driven by application issues. This is to be expected, but it has resulted in structure that

is not as efficient as expected because of a sensitivity to field damage and fatigue degradation driven by the dominant fatigue damage process (i.e., matrix cracking and delamination). Potential for structural efficiency improvement may mean that the ultimate payoff from life prediction methodology is the ability to design while relying to a greater degree on the inherent strength of composite materials. Emerging material forms (i.e., three-dimensional preforms, pultrusions, braids, and knit multiple orientations), together with recognition of microscale, ply thickness, and laminate stacking order as critical design issues, compound the problem while becoming a necessary part of the solution. A new design and analysis paradigm, based on a physical understanding of the failure processes and in a format amenable to the designer, must be available in real time to guide the development of a specific design. The importance and microscale details and attendant new material forms preclude the new paradigm from being based on metals technology.

3

CONSTITUENTS

Failure mechanisms in fiber composites depend upon the characteristics of the fibers, the matrix, their interface or interphase, and the microstructure and macrostructure of the material. The macrostructure includes the laminate stacking sequence and the interply geometry and resin content. The microstructure includes the fiber volume fraction and its distribution uniformity, the fiber waviness, and, in the case of braided or woven fiber composites, the fiber weave pattern. While the microstructural arrangement of the constituents is of first-order importance, it is not the subject of this chapter. The mode of failure also strongly depends upon the orientation and direction of the applied loads or displacements. Unlike the case of static loads, there is little literature on the analysis of relationships between fatigue properties of unidirectional fiber composites and those of its constituents. The fatigue properties of such a composite should probably be considered to be defined experimentally, but an understanding of the important mechanisms of failure initiation is still required to understand the process of damage growth as it relates to failure of a laminate.

It appears that the fatigue behavior of a unidirectional fiber composite differs in a fundamental way from that of a metal. Metals are polycrystalline aggregates. During stress cycling, microdamage develops in the form of microcracks, void growth, and plastifications of single crystals. At some state of the cycling, a dominant crack develops and grows with continued cycling until the specimen fails. These two stages are known as *initiation* and *propagation*. In the propagation stage, the growth of the crack with cycling can be predicted reasonably well. The use of a crack growth law, with a measured or assumed size of the existing crack, forms the basis of design methods for the fatigue of metals. The methods take into account possible crack locations and characteristics of available inspection methods to define safe inspection periods.

For the case of a unidirectional fiber composite, however, the nature of damage initiation and growth is quite different. The microstructure consists of a polymeric matrix containing stiff, strong fibers. The material is very anisotropic, and cracks propagate easily in the matrix along surfaces parallel to the fibers. When a unidirectional specimen is cycled in tension in the fiber direction, damage accumulates in the form of random fiber ruptures that are the source of small cracks in the matrix along the fibers. More and more cracks develop with additional cycling until some coalesce to produce catastrophic failure. The failure surface is jagged and irregular, and the failure mode is similar to the static tensile fiber mode. The entire lifetime is spent in a damage initiation phase.

Failure modes for other load conditions may be expected to be distinct and complex. The failure criteria for combined cyclic stress must therefore treat physically realistic failure modes. These issues are discussed in this chapter in the context of the lifetime characteristics of the constituents of a polymer matrix composite.

MATRIX MATERIALS

This section will discuss the mechanical and physical properties of the most common polymer matrix materials that are used with reinforcing fibers. Resins (also called polymers and plastics) are generally divided into two main classes: thermosets and thermoplastics. The general characteristics of these types of resins, and their use in composites, are also examined.

Polymer matrix thermoset composites generally have low transverse tensile, through-the-thickness, and in-plane shear failure strains. For currently used material forms, the failure strains are 0.4 to 0.5 percent. Under stress, transverse microcracks will occur in the matrix phase. This low threshold of microcracks is often attributed to the brittleness of the matrix. The problem of formation of microcracks in a composite has been researched from the point of view of matrix properties effects on composite performance. That body of data is important and is reviewed below to facilitate judgment of the range of cracking characteristics that can be produced by means of matrix selection.

Lifetime Trends of Thermoset Matrices

Traditional thermoset polymer matrices are relatively brittle and, like most brittle metals and ceramics, are not very sensitive to cyclic fatigue loading (Mandell, 1990; Hertzberg and Manson, 1980). Paris law fatigue crack growth exponents (based on the Mode I stress intensity factor) range up to 20 or more depending on the degree of cross-linking and other factors. Tensile fatigue S-N curves are difficult to obtain and interpret for brittle systems; lifetime trends from S-N data have relatively low slopes and appear to be dominated by crack nucleation with apparently high fatigue limits (Odom and Adams, 1983; Mandell, 1990). The primary environmental factor that has been explored is moisture, which appears to have little effect on fatigue beyond the generally reversible decrease in static properties associated with swelling (Mandell, 1990).

In an effort to minimize resin microcracking and increase delamination fracture toughness of a composite, much work has been done to toughen neat polymer matrices. Clear distinction must be made between a tough neat polymer matrix itself and a tough composite, however. Toughness is generally defined as high energy absorption or high fracture resistance. For a two-phase (rubber or thermoplastic particle toughened) matrix system, the proposed energy dissipation mechanism includes elastic cavitation, shear yielding of the matrix (Kinloch et al., 1983a, 1983b), and tearing of the elastic particles that bridge the crack (Kunz-Douglass et al., 1980; Kunz and Beaumont, 1981).

Toughening of a matrix with a discrete second polymer phase has a history of about 20 years. Polymers often used in epoxies include CTBN rubber (carboxyl terminated butadiene-acrylonitrile copolymer) and thermoplastics such as blends of polyethersulfone and polyetherimide or phenol and amine functionally terminated polysulfone oligomers (Bauer, 1986). It is generally accepted that this approach does indeed improve fracture resistance, impact tolerance, and some shear properties without substantially degrading the thermal and mechanical properties of the neat epoxy and its composites. However, confusion and controversy do exist, mainly centering on the degree and price of improvement. This is understandable because the degree to which an epoxy can be toughened depends on many factors. The key ones are the base epoxy type and the molecular weight between cross-links, the second-phase material and size distribution, the curing agent and its compatibility with other components, and the curing conditions. With regard to the toughness of composites, another key factor is the spacing available between adjacent fiber filaments and between plies. As a rule, the spacing available is considerably smaller than the critical plastic zone diameter of the crack tip for the toughened epoxies (Bascom et al., 1980).

Clearly, this geometric constraint in a composite considerably reduces the effectiveness of the toughened matrix. In summary, the status of toughened epoxies is as follows:

- For the very brittle epoxies (DGMDA/DDA, $G_{1c} \approx 80$ J/m^2), the degree to which one can be toughened is very limited, because the rubber toughening approach is ineffective, perhaps due to the high cross-link structure of the epoxy. In addition, this approach has more problems due to chemical interactions in the "hot/wet" environment. Fortunately, composites made from brittle epoxy generally nave a G_{1c} that is at least a factor of two higher than the neat resin itself.

- For tougher epoxies (bis-phenol A, $G_{1c} \approx 270$ J/m^2), the two-phase toughening approach is effective. It can increase G_{1c} from severalfold to 10-, 20-, or even 50-fold depending on the formulation and the molecular weight between cross-links. The penalty paid for this approach is reduced service temperature and reduced stiffness of the matrix. However, if the CTBN rubber addition is less than 10 percent to a bis-A epoxy system, the temperature reduction is quite moderate while still realizing a severalfold increase of G_{1c}.

- In composites containing toughened epoxies, the improvement in G_{1c} is clearly evident. The degree of improvement, however, is in dispute and is probably caused by the many variables involved in testing toughness.

The key drawback of a highly toughened epoxy matrix is the attendant reduction in service temperature. At present, a system with acceptable toughness is limited to about 120°C due to the second (toughening) phase effects on matrix strength and stiffness at elevated temperatures.

New Epoxy Development

The multitude of requirements (i.e., high temperature, high toughness, low creep, and low moisture absorption) for a high-performance composite cannot be completely met by conventional epoxy technology. In fact, this is not possible based on the epoxies discussed so far. The chemical structures are such that increased cross-linking results in decreased toughness and toughenability (Yee, 1984). However, increasing the cross-link density of these systems does increase the temperature resistance, which unfortunately increases chemical interactions with water as well when diamines such as DDS are used as curing agents because of the increased hydroxyl group density. It should be clear, then, that these requirements are fundamentally in conflict and mutually exclusive as far as conventional epoxy technology is concerned.

Recent work (Schultz et al., 1988) on a new class of epoxy and curing agents based on tricylic hydrocarbon, called fluorene, as the backbone structure looks promising, however. The fluorene compounds are inherently high temperature resistant and have a low water-sensitivity, nonpolar structure. The curing mechanism uses a mixture of two hardeners, with one controlling chain extension to build a linear flexible structure and the other controlling cross-linking. Basically, one has the choice of any desired cross-link density and, hence, toughness. The toughenability of these resins with rubber particles and their blendability with the bis-phenol A resins have also been demonstrated by the researchers. There also appears to be the potential to have an epoxy system that can be processed by conventional method with properties better than that of the TGMDA/DDS system in every aspect: 20+°C in-service temperature, less than 1/3 of the water absorption, and easily 10 times more fracture energy.

Lifetime Trends of Thermoplastic Matrices

Many high-performance thermoplastic polymer matrices are attractive because of their high fracture toughness. However, their toughness advantage over thermosets may be at least partially offset by their increased sensitivity to sustained or cyclic loading, as well as environmental stress cracking in certain cases. Most thermoplastics show both a Paris law fatigue crack growth exponent on the order of four (based on the stress intensity factor) and a greater fatigue sensitivity than thermosets (Hertzberg and Manson, 1980; Mandell, 1990). This fatigue sensitivity is similar to that of many aluminum alloys. While most available delamination fatigue crack growth data for composites with thermoplastic matrices show higher exponents than for the neat matrix (Mandell, 1990), some data for carbon fibers have been reported with the equivalently low exponent for reversed Mode II loading (Russell and Street, 1987; reported as an exponent of 2 on the strain energy release rate, which is equivalent). Environmental stress cracking has also been a major problem for thermoplastics in specific environments and must be carefully considered in each case (Kinloch and Young, 1983). Transitions to brittle environmental stress cracking may only be observed after very long loading times and at low stresses, making real-time studies difficult (Kramer, 1979). In neat thermoplastics, both the fatigue and environmental sensitivities derive primarily from effects on fibrous craze structures at crack tips, which are not usually observed in thermosets (Kinloch and Young, 1983).

Thermoplastics generally are tough compared to thermosets and are widely used without reinforcement. However, their stiffness and strength properties, although similar to those of thermosets, are low compared to other structural materials, so use of reinforcements is desirable. Thermoplastics can be formed into complex shapes easily and economically by processes such as injection molding, extrusion, and thermoforming. The creep resistance of any thermoplastic, particularly at elevated temperature, is significantly lower than that of thermosets, and this has been a serious impediment to their wider use in structural applications. As a class of materials, thermoplastics are also more susceptible to attack by solvents than are thermosets.

To date, thermoplastics have been reinforced primarily with discontinuous glass fibers and particulate fillers; there has been relatively little work with continuous fibers, with the exception being in the area of relatively high temperature resins like polysulfones and thermoplastic polyimides. The elastic and strength properties obtained with composites using these matrices are similar to those employing epoxies, and their impact resistance is better. Although they are harder to process than epoxies, the good moisture resistance and high temperature range of thermoplastic polyimides make them attractive. Polysulfone composites are easier to fabricate than thermoplastic polyimides, but their relatively poor chemical resistance to solvents is a severe limitation.

It appears likely that high-performance thermoplastic-matrix composites can become an important class of high-performance structural materials with additional development. High-performance thermoplastics of recent vintage seem to offer service temperature, mechanical properties, and environmental resistance from common solvents generally comparable to those of epoxies for fiber composites. Presently, there are about a dozen polymers that can meet the above description. They are basically in two groups: semicrystalline polymers and amorphous polymers. Several points can be made about current high-performance thermoplastics:

- Rated service temperature (probably ambient RH) for high-performance thermoplastics ranges between 170° and 250°C. This is generally higher than that of most epoxies, including the TGMDA/DDS system; fluorene epoxies may be comparable.

- Composite compressive strength is generally poor, which is troubling. Poor compression strength of thermoplastic-matrix composites is most likely due to their lower shear modulus than highly cross-linked epoxies. Poor interfacial bonding may also be partly responsible in some resins.

- Neat thermoplastics with the exception of PPS are about a factor of 10 higher in terms of matrix toughness, G_{1c}, than a brittle epoxy. However, this difference is considerably reduced when translated to graphite composites, because the toughness of a composite with a brittle matrix increases to severalfold higher than the toughness of the matrix itself, and the toughness of a composite with a tough matrix decreases to a level lower than the matrix toughness.

- Graphite-thermoplastic composites seem adequate in toughness in a static test, but there is some indication that the degradation of matrix G_{1c} is much faster under fatigue loading than the toughened epoxies, which are in turn faster than the brittle epoxies (O'Brien, 1986).

- Key concerns about thermoplastics when considering the lifetime of composites are the time-dependent behaviors of thermoplastics (e.g., creep), the definitely nonlinear stress-strain relationships, the question of the fiber-matrix bond in composites, and their generally poor compression strength.

Prepreg processing is also key in cost considerations. For amorphous polymers, the solvent coating technique for making prepregs (whether it is a dissolved polymer or in situ monomer reactions in a solvent) seems to be the method of choice. For semicrystalline polymers (there is hardly a known solvent), the hot-melt method has dominated at present. Other methods such as film stacking, powder coating, and fiber comingling are being studied. All processing methods for lamination also involve very high temperature (300°-430°C), vacuum pressure, and time for working the voids. A further complication is the necessity to develop, control, and maintain the degree of crystallinity in this group of polymers in order to have consistent quality of the fabricated composites. Volatile removal and composite consolidation are difficult tasks. To date, there is no indication that thermoplastic composites will ever realize the strongly anticipated cost advantage over epoxy composites. However, recent advances in resin transfer molding techniques may lead to substantial improvements in fabrication speed and economics. For example, the development of ring opening catalysts for bis-phenol A cyclic oligomers allows the use of low-viscosity compounds for impregnation of fiber preforms followed by ring opening, which results in a high molecular weight, high-viscosity matrix polymer. Clearly, the development of new chemical compounds and/or polymerization techniques may dramatically alter current concepts regarding the processing economics and technical difficulties associated with both thermoplastic and thermoset matrix materials.

FIBERS

There are several major types of man-made reinforcing fibers, including glass, graphite (carbon), organic, boron, and ceramic. This section considers the properties of these fibers and emphasizes the most important (i.e., glass, graphite, and organic).

The major reinforcing fibers have tensile stress-strain curves that are linear to failure. These fibers are sensitive to imperfections, which have two important effects on tensile strength: there is considerable scatter at a given gauge length, and the mean strength decreases with increasing gauge length. Therefore, a reported value of mean fiber strength is useless unless the associated test length is specified. Furthermore, the amount of scatter is important in evaluating fiber strength. Ideally, fiber strength properties should be measured at several lengths.

The problem of understanding fiber tensile strength is complicated by the fact that it is measured using single fibers, untwisted bundles of fibers (ends), impregnated ends, twisted ends (yarns), and composite tensile coupons. As a rule, each one gives a different strength value. Fiber strength values as reported here are based on single-filament tests and unidirectional composite coupon tests. The latter data are of greatest relevance to designers.

Glass Fibers

Glass fibers are the cheapest and most widely used man-made composite reinforcements. They are also the oldest, dating back to the period of World War II. Glass fibers generally have high strength-to-weight ratios, but their elastic moduli, which are in the range of those of aluminum alloys, are low compared to the newer fibers such as graphite and aramid. The internal structure of glass fibers is amorphous (noncrystalline), and they are generally regarded to be isotropic. As fibers comprise only a part of the composite, and as they are generally oriented in several directions, the modulus-to-density ratios of glass-fiber-reinforced plastics are substantially lower than those of metals. This is one of their major limitations as a structural material and is directly related to their inferior acoustic properties.

The creep resistance of glass fibers at room temperature is substantially better than that of plastics, but not as good as that of structural metals like aluminum and steel. The addition of glass fibers to plastics greatly reduces their tendency to creep. As in the case of creep, the properties of glass at room temperature fall between those of plastics and metals, but are much closer to the latter.

The coefficient of thermal expansion of glass is an order of magnitude lower than that of most plastics and is lower than most aluminum and steel alloys. Reduction of the thermal expansion coefficient is an important reason for the addition of glass fibers to plastics for many applications. A material with a low coefficient of expansion is frequently said to be dimensionally stable, but other factors enter into resistance to dimensional changes, including creep and swelling from moisture absorption. Glass fibers generally have good chemical resistance and are noncombustible. They do not absorb water, but their tensile strength is substantially reduced chemically by means of stress corrosion cracking in the presence of moisture. The strength, modulus, and creep rupture resistance of glass fibers decrease with increasing temperature. Conversely, creep rate increases. However, the useful temperature range is quite large. Glass does not soften substantially until temperatures over 500°C are reached.

The most widely used reinforcing fiber, by far, is E-glass, where the E designates electrical grade. This lime-alumina-borosilicate glass does not have a fixed composition. Producers vary the constituents based on raw material costs and process considerations. Within prescribed ranges, variations in glass formulation are not thought to affect mechanical properties substantially.

S-glass is a high-strength glass initially developed for military applications. Its modulus is about 20 percent greater than that of E-glass and it is about one-third stronger. The failure energy of S-glass fibers is high, and the impact resistance of composites made from them is among the highest of all fiber-reinforced materials. Despite its generally better properties, the use of S-glass is far more limited than E-glass because of its higher cost.

Individual E- and S-glass fibers show similar well-established lifetime trends and environmental sensitivities. Inorganic glass materials in general are not sensitive to cyclic stressing (Evans, 1980; Mandell et al., 1985), but they do show a cumulative time under load sensitivity termed static fatigue (in practical terms this is the same as creep rupture, but it is not due to a

creep mechanism). The static fatigue effect is caused by the stress corrosion growth of surface flaws on the fiber and occurs in the presence of even the minute quantities of moisture usually present in composites (Mandell and Meier, 1983). The stress corrosion effect in E-glass results in a moderate time sensitivity under static or cyclic loads of about 3 to 4 percent of the short time strength per decade of cumulative time extending from very short to very long time scales (Mandell and Meier, 1983). Much more severe stress corrosion of glass fibers and composites is evidenced under even dilute acid solutions (Noble et al., 1983), which has led to many service failures (Hull and Hogg, 1980).

The static fatigue behavior of glass fibers and strands is consistent with that of bulk glass (Aveston et al., 1980), but a significant cyclic fatigue sensitivity not present in bulk glass or individual fibers becomes evident at the multifilament strand level, whether dry or impregnated (Mandell, 1982; Mandell et al., 1985). The cyclic tensile fatigue effect, apparently due to fiber-fiber interactions, produces a stress versus log cycles fatigue lifetime trend of about 10 percent of the one cycle strength per decade of cycles, which is also observed in most glass-fiber-dominated constructions under most tensile fatigue conditions (Mandell, 1982, 1990). No clear fatigue limit can usually be defined. It appears that small impregnated strand samples can adequately define the macroscale composite behavior, if the failure is fiber dominated. However, woven fabric constructions can show much greater cyclic fatigue sensitivity than strands in some cases (Mandell, 1975), as indicated in Figure 3.1. A corresponding decrease in the fatigue crack growth exponent is observed with woven fabric reinforcement for in-plane crack growth that causes tensile fiber failures (Mandell, 1975, 1982). The greater fatigue sensitivity of woven fabric reinforcement is associated with weave cross-over delaminations (Mandell, 1975).

Carbon (Graphite) Fibers

Carbon fibers comprise one of the most important classes of reinforcement, with enormous potential for future growth. Their primary advantages over glass fibers are higher modulus, lower density, better fatigue properties, improved creep rupture resistance, and lower coefficient of thermal expansion. Creep at room temperature is generally considered to be negligible. On the negative side, the high moduli of carbon fibers cause their failure energies to be relatively low. As a result, the impact resistance of graphite fiber composites is generally lower than that of glass fiber composites. It should be emphasized, however, that the subject of impact resistance is very complex and generalizations can sometimes be misleading.

Graphite fibers of commercial importance are currently made from three precursors: polyacrylonitrile (PAN) fiber, rayon fiber, and petroleum pitch, which is a residue of the refining process. Production of high-modulus graphite fibers by pyrolysis

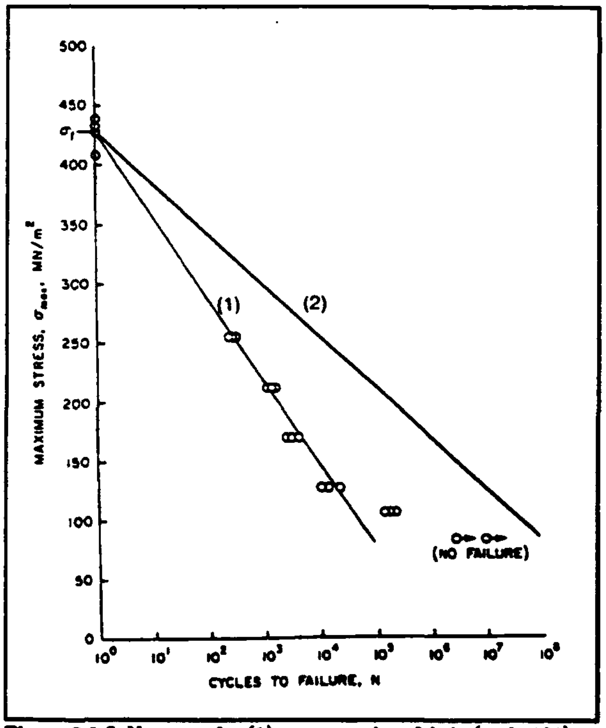

Figure 3.1 S-N curves for (1) a woven glass fabric (style 181) reinforced polyester (Mandell, 1975) and (2) a nonwoven glass fiber composite in tension (R = 0.1).

of rayon and PAN fibers dates back to the 1960s (low-modulus carbon fibers produced by pyrolysis of rayon cloth and used for reentry vehicles and rocket nozzles were introduced in the 1950s). Spinning of graphite fibers from pitch is a more recent development. To date, pitch fibers have been able to match the moduli that can be obtained with PAN- and rayon-based fibers, but their strength properties are substantially lower. The high strength and moduli of graphite fibers, regardless of precursor, result from their high degree of crystallinity and orientation. Graphite fibers are strongly anisotropic because of their high degree of internal structure orientation. Transverse extensional modulus and shear moduli for most fibers are generally an order of magnitude lower than axial modulus.

The most important graphite fibers at the present time are PAN based. There are many graphite fibers on the market. They can be divided, somewhat arbitrarily, into three major categories: high strength, high modulus, and ultrahigh modulus.

Relatively little study has been made of fatigue lifetime trends of carbon fibers and strands. Results on model composites (Lorenzo and Hahn, 1986) suggest that the trends of fatigue life of unidirectional impregnated tows and unidirectional coupons represent fiber-dominated behavior as with glass, with failures originating at individual broken fibers; however, this point remains poorly established. Carbon fibers are very resistant to creep rupture below temperatures where oxidation takes place. Carbon fibers are among the most fatigue-resistant materials known. They are also very resistant to most environmental agents such as moisture, but can take part in electrochemical reactions as with aluminum.

Organic Fibers

Some natural organic fibers, such as cotton, jute, and sisal, are also used as reinforcements. However, their mechanical properties are modest because of their low modulus, and they are of little interest for structural applications. The same is true of synthetic organic fibers, with the exception of aromatic polyamides (aramids).

There are several commercial aramid fibers: Nomex Kevlar, Kevlar 29, Kevlar 49, Technora (by Teijin), and Kevlar 129. Applications of Nomex include high-temperature fabrics, filters, and structural honeycomb for sandwich-core laminates. Kevlar is used for tire cord. The two fibers of interest for reinforcing plastics are Kevlar 49 and, to a lesser extent, Kevlar 29.

Kevlar 49 fibers have excellent tensile-tensile fatigue resistance. Their compressive fatigue characteristics have not been studied extensively, but they appear to be good within the proportional limit, which is very low. The resistance to creep of Kevlar 49 fibers is significantly better than that of other organic fibers, but room temperature creep rates as a function of stress are of the order of magnitude as those of glass and should be considered in design. The creep rupture resistance of Kevlar 49 falls between that of S-glass and that of graphite.

Kevlar 49 fibers are strongly anisotropic, a property they share with graphite. Transverse extensional modulus and shear moduli are about an order of magnitude lower than axial extensional modulus. The tensile energy to failure of Kevlar 49 is much greater than that of high-strength graphite, about the same as E-glass and significantly lower than S-glass. Although the area under the compressive stress-strain curve of Kevlar 49 is very large, the energy to the proportional limit is relatively low compared to other fibers. These and other factors cause the impact behavior of composites reinforced with Kevlar 49 to be particularly complex. Generally, the impact resistance of these materials falls between those of E-glass and high-strength graphite composites, however.

Kevlar 49 fibers have been studied under both creep rupture and cyclic loading conditions (Kenney et al., 1985). The results indicate a stress versus log time slope of about 3 percent per decade for individual fibers under both static and cyclic loading. Extensive impregnated strand creep rupture data (Chiao and Chiao, 1982) show a general trend of about 5 to 6 percent of short time strength per decade of time under load, similar to that of unidirectional material under cyclic loading (Roylance and Roylance, 1981). Thus, it appears that the individual fibers tend to fail by a strain-limited creep rupture process, as do many oriented organic fibers (Kenney et al., 1985), with some effects of moisture pickup in the fibers (Roylance and Roylance, 1981).

INTERFACE-INTERPHASE ZONES

While several direct test methods for determining fiber-matrix bond strength have been developed in recent years (as reviewed in Narkis et al., 1988), directly measured lifetime trend data are not available. The best available data are from observations of debond initiation at coupon edges during fatigue loading (Owen and Bishop, 1973; Mandell, 1990). These data, all for thermoset matrices, tend to show somewhat greater fatigue sensitivity for debonding than for the neat resin S-N data (Mandell, 1990). However, considering that debonding is usually the first damage event observed in fatigue, very little is known about its durability in a fundamental materials sense.

The interface also plays a key, yet obscure, role in long-term environmental resistance. For thermoset systems under hot/wet conditions, any irreversible damage (in the absence of mechanical loading) has usually been traced to bond failure (Hancox, 1981).

The presence or absence of interfacial bonding between the fibers and matrix also affects the residual stress state that occurs due to the difference in thermal expansion coefficients for the fiber and matrix. The absence of an interfacial bond precludes the development of tensile stresses across the fiber-matrix interface, but has no effect on compressive stresses. Stress transfer across an interface is therefore dependent on the residual thermal stress field and the superposed stress field resulting from external loading.

The relatively random location of fibers precludes an exact analysis of the thermal stress field. While models incorporating either periodic fiber placements or single fibers surrounded by a concentric shell of matrix give estimates of the mean stresses that can be expected due to thermal expansion mismatch, the results are misleading. It has been shown (S. Sternstein, presentation to committee) that random placement of one or two fibers in an otherwise periodic hexagonal array of fibers can alter the radial stress component at an interface from compressive to tensile. Such calculations strongly suggest that the interface plays a major role in determining both the residual thermal stress field and the value of external loading that will result in an interfacial radial tensile stress sufficient to cause interface separation if not bonded, or interface failure if bonded.

It is safe to assume that the lack of interfacial bonding will be deleterious to all failure properties that involve stress transfer from fiber to fiber through the matrix. Delamination strength and fracture toughness of the composite and transverse tensile strength are most likely to be reduced by poor interfacial bonding, although the degree of reduction may be partially offset by large compressive residual thermal stresses when present.

4

UNIDIRECTIONAL COMPOSITES

The unidirectional composite is the basic form of the material and is generally studied to provide data for the design of composite laminates. This procedure is well defined and widely utilized for such nondestructive properties as expansion coefficients and stiffness. However, the relationship between unidirectional materials and laminates for destructive properties is not well defined, and a number of general approaches exist to address this need. At the present time, there is growing use of laminate data to derive the strength and life characteristics of a unidirectional composite. With this approach, the unidirectional composite is treated largely as an analytical abstraction that facilitates laminate design, rather than as a practical engineering entity. This current status is the result of experimental problems introduced by a combination of large strength and stiffness ratios parallel and perpendicular to the fibers and in shear, that give rise to extreme experimental difficulties. However, unidirectional composite strength and lifetime failure processes can be observed by means of appropriate coupon and laminate testing and analysis. The need here is to isolate the critical local failure mechanisms and characterize them. This requires definition of the representative volume element for each failure mode. The critical elements identified for current polymer-matrix materials are fiber tensile failure, matrix cracking, interlaminar debonding, and local compressive instability.

FIBER BUNDLE TENSILE FAILURE

The tensile fiber failure representative volume element is defined by a unidirectional composite element where stress is applied in the direction of the fiber axis. This is the dominant strength and stiffness axis for composites, and the first-order characteristics of this representative volume element are well developed.

Tensile Fiber Strength

A composite structure can exhibit extremely high static strength to weight when the failure mode exhibited is brittle tensile fracture that cleaves fibers. The structural efficiency of composites in this mode provides the impetus for use of the materials. While coalesced-through-thickness cracks do not naturally exist in composites, a single dominant failure surface can form just prior to fracture and cleave the composite. This surface is generally a coalescing of dispersed damage zones rather than the propagation of a crack (Rosen, 1964). When the composite has large-diameter fibers with small coefficient of variability of strength, a crack may propagate due to the dynamic energy released. This propagation may also occur when there are boundary stress concentrations due to structural discontinuities. This fracture behavior has been confirmed in experiments by means of arresting and examining dynamic fractures in composite structure (Eisenmann and Kaminski, 1972). This fracture behavior is triggered from laminate boundary

stress concentration, induced by planned stress concentration or service damage. Massive damage such as brooming failures may be the result of dynamic effects after initial propagation from a stress concentration or may result from widely dispersed damage in the composite material.

The role of the process zone in the fracture of composite structure was defined by Waddoups et al. (1971). This work demonstrated that classical Griffith fracture criteria could be applied to the analysis of fracture of composite material by means of correcting the physical flaw size with an estimate of the process zone scale. This approach provided the foundations for treating stress concentrations of arbitrary size and shape. These ideas were later extended to include any analysis of the process zone using a Dugdale-Barenblatt model for the process zone effects (Backlund and Aronsson, 1986; Aronsson and Backlund, 1986). The authors were able to compute accurate and effective stress concentrations for shapes ranging from round holes to square holes to cracks. In addition to the computation of effective stress concentration, the authors were also able to estimate the process zone dimensions to be on the order of 2.0 to 4.0 mm. Extension of these concepts to the analysis of complex elements, such as bolted joints, has been done (Eisenmann, 1976), but the computer codes for bolted joint analysis remain proprietary property.

The very small unidirectional tension process zone defines the appropriate representative volume element for this mode. The apparent strength of a composite structure is statistically complicated by the small size of the process zone as the probability of rupture is scaled to the size of typical structures. The nature of the scaling depends upon the mode of failure. In structures with large stress concentrations, the location of initial failure is predetermined, and failure of the first representative volume element can lead to a propagation-type failure. Here, a weakest link failure criterion is appropriate and size effect is very significant. In the absence of important stress concentrations (as in the case of local reinforcement), failure will result from an accumulation of dispersed local damages and a chain-of-bundles model will represent the size effect, which will be much less severe.

Micromechanical analyses can predict first-order size effects in the strength of composite materials (Rosen, 1964; Phoenix and Sexsmith, 1972); subsequent tests experimentally corroborated those predictions (Waddoups et al., 1971). Bullock (1974) showed that three different specimen types, each of which fail in fiber tension, exhibited size and shape scaling effects that could be correlated by using the results of Weibull (1939). Those specimens were drawn from rigorously specified T300/5208 graphite epoxy. To the first order, the specimens were found to have constant shape parameters and strength in proportion to the statistically weighted stressed volume (now correctly understood as a length effect). As a result of these experiments, it has been shown that the similar failure processes yield consistent Weibull shape parameters. Over long production runs, again using rigorously specified T300/5208 graphite epoxy, Weibull shape parameters as high as 20 (a coefficient of variation of about 5 percent) have been observed for this representative volume element.

The shape parameter of the fracture for bearing failure in a bolt-loaded hole (with fibers at $\pm 45^\circ$ to the load axis) is of the same order as the shape parameter for the simple tensile and flexure specimens (Wolff and Wilkins, 1980). In the case of the bolt-loaded hole, suspended tests positively showed the development of a fiber cutting process zone at the fiber tangent point. Further experiments demonstrate that large-scale box beams, fabricated from T300/5208 graphite epoxy, failed statically in the same brittle fracture mode as the coupons (Wolff and Wilkins, 1980). The scatter of the three full-scale tests was small enough to conclude that they could have likely been drawn from the same population as the small-scale specimens and coupons.

As a result of these observations, a shape- and scale-independent point stress or strain-based failure model cannot properly predict composite strength when fibers are cleaved by the fracture surface. While the Weibull shape parameter is invariant to scale, the expected strength (such as stress at failure) is a function of the scale of the structure. The process zone (and subsequent unidirectional tension representative volume element) that develops when fibers are torn is so small that a minimal representative specimen (such as a tow or a flexure specimen) has sometimes been used to develop the required data to characterize expected strength and variability at greatly reduced test cost.

Tensile Fiber Fatigue

The fatigue behavior of fibers in tension was presented in Chapter 3. The fatigue sensitivity of the fiber directly translates to the representative volume element only when matrix cracks or fiber-matrix interface failures are not present. For brittle matrix materials, or for viscoelastic matrix materials that cause stress redistribution, the fiber bundle tensile representative volume element lifetime will differ from the fibers themselves. In either case, the fatigue insensitivity of fibers such as boron and graphite leads to a nearly immeasurably small strength loss per decade under either constant-amplitude or random-amplitude loading. Typical power law exponents as high as 20 to 30 are observed for this representative volume element. As shown in Figure 4.1, even when transverse plies are in parallel with the fiber bundle, the tension fatigue life of the composite is significantly better than aluminum. When appropriate design details subdue competing failure modes that progress at more rapid rates, structures fabricated from fatigue-resistant fibers can exhibit nearly indefinite fatigue life.

MATRIX CRACKING

Ninety degree matrix cracking was observed in boron epoxy in situ in a 0/90° laminate (Waddoups, 1968). These observations led to the formulation of a matrix cracking limit for design purposes in the form of a transverse strain limit. This strain limit was maximized by means of material design by controlling resin content and transverse fiber spacing. The transverse cracking is a life-limiting failure mode for unidirectional, off-axis systems such as graphite-polymer systems where the microgeometry is not as controllable as in boron epoxy.

Transverse Strength and Fracture

Isolated tests of transverse properties are very difficult to conduct because of low strength, low toughness, and edge-defect sensitivity of the pure 90° laminate. In tests on this orientation, the transverse response is nonlinear, time dependent, defect sensitive, and irreversible due to crack and microcrack formation. The transverse strength distribution for T300/5208 has a Weibull shape parameter of 10 or less. Fortunately, this failure process does not dominate static strength in a well-optimized structure. As demonstrated by Parvizi et al. (1978), the onset of transverse cracking is a first-order function of ply thickness. The impact of ply thickness on transverse cracking has also been examined by Lagace and Nolet (1986) and Nairn (1989).

The onset of transverse cracking has been demonstrated to be an instability phenomenon, with energetics models providing excellent correlation for the ply scaling effects. In glass-epoxy systems, thin plies (less than 0.13 mm) can extend the cracking threshold in transverse plies beyond half of the strain to failure of the unidirectional material. Manufacturing trends toward thick plies for low-cost construction must be balanced against expected life requirements.

The use of point stress or strain laws to describe the expected transverse strength is not appropriate for composites. There are two critical effects. First, multiple transverse ply cracks are expected prior to laminate failure. Second, scale not only influences longitudinal strength, but also transverse strength. Various mechanics models predict multiple ply cracks. Energetics models to predict dense cracking thresholds are sensitive to the ply geometric constraint and may be preferable for design calculations.

Transverse Fatigue

Fiber placement in a composite is random in the transverse plane; consequently, dispersed transverse damage will propagate at low levels of applied stress from touching fibers and other defect sites. Initial transverse matrix cracks in a laminate can actually relieve stress, while coalescing transverse matrix cracks can produce delamination and eventual laminate failure. The advancement of transverse matrix flaw fields can be adequately modeled with a power law growth model.

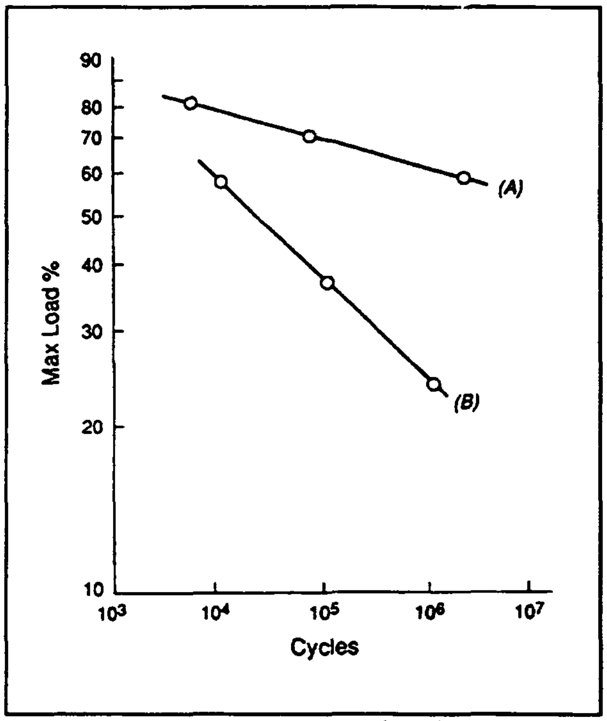

Figure 4.1 Tension fatigue of composite (A; $\pi/4$ graphite epoxy; log slope= -18.5) versus aluminum (B; 2024 T_3 aluminum; log slope= -5.2).

INTERLAMINAR DEBONDING

Typically, it is very difficult to design a composite structure that does not subject the laminate to out-of-plane loads. Delaminations typically initiate at free edges, at intersections of microcrack fields, at lamination boundaries, at intentional internal defects such as ply dropoffs or bonded joints, and at manufacturing defects such as trapped gas interfaces or near foreign materials. Each of these initiation sites has accompanying out-of-plane loads to propagate flaws at subcritical loads. Solution of the free edge problem demonstrated that lamination details alone provide sufficient out-of-plane driving force to influence fracture of a laminate (Pagano and Pipes, 1971).

The delamination representative volume element is fundamental to understanding damage growth in a composite structure. The delamination representative volume element may vary in size from the scale of the fibers to square meters. In engineering structures, energetics-based analyses have been used successfully to correlate delamination initiation, stability, and subcritical growth. This high degree of correlation of the physics of delamination growth removes the need to discuss the strength and life of this representative volume element as separate topics.

Typically, the onset of delamination instability is coincident with Weibull shape parameters of 10 or less. Delamination growth, as measured from S-N data by means of logarithmic slope or directly from fracture testing, is bounded by a release rate exponent of 5 or less, even for brittle epoxy-based laminates. These relatively rapid growth rates, coupled with the

potential stress concentration or instability resulting from the growth, make this representative volume element critical for composite structure. Fatigue tests ranging from the scale of coupons (S. S. Wang, 1981), open holes in the interior of coupons (Reddy et al., 1987), and large-scale box beams containing mechanical joints (Wolff and Wilkins, 1980) have all ultimately failed due to stress concentration or instability resulting from delamination growth.

The risk of delamination as a failure mode is offset by demonstration of the predictability of delamination growth with first-order energetics-based models. Successful integration of delamination growth on a cycle-by-cycle basis given a flight service load spectrum is reported by Wilkins et al. (1982). For the F-16 horizontal tail components (Wilkins, 1983a), delamination control and full-scale tests demonstrating both stability and detectability of delaminations formed the basis for execution of a U.S. Air Force-specified damage tolerance program. Understanding the relationships of the propensity of delamination growth to design detail is essential for the design of fatigue-resistant composite structure.

LOCAL COMPRESSIVE INSTABILITY

As in the case of tension fatigue, local compressive instability may be regarded as the result of an accumulation of dispersed damage sites that grow in size and number. The difference is that the local failure of damage sites is the result of different mechanisms for compression than they are for tension. There are both experimental and analytical indications that the beginning of the failure process in a unidirectional composite subjected to compression in the fiber direction is likely to be a local instability. Stress-induced local material strength failures and excessive local deformation may also occur.

If all fibers are perfectly oriented and if the dominant stress component is compression in the fiber direction, the expected failure mechanism is instability of the material. This can be followed by stress redistribution because of the buckling and by failure in some other mode such as fiber breakage during bending, leading to failure of the composite as a whole. There is also the possibility that there can be a fiber material strength limitation. For example, a Kevlar fiber is known to have low compressive strength due to microinstability of the oriented polymer chains in the fiber.

The practical material is much more complex because there are imperfections of many types within the material. The imperfections that create secondary stresses may be fiber misalignments, curvatures, disbonds, inclusions, voids, or microcracks. All perturb the basic stress field and have one of two effects: they can be large enough that by themselves they cause a matrix-dominated failure mode prior to any material instability or, if that does not arise, instability will occur at loads modified by the presence of these secondary stresses within the material. Over the years, various investigators have postulated different types of secondary stress influences and have generated mathematical models to explain this. Most have shown correlations between the mathematical models and experimental results. There are numerous analytical models whose results agree with experiments. However, there is still a strong lack of understanding of the failure mechanisms of these materials.

Important effects on compressive strength result from changes in fiber orientation, fiber volume fracture, and interface conditions--parameters that are not well controlled during the fabrication process. Therefore, it is reasonable to believe that the material to be analyzed varies from point to point throughout its volume. Hence, a proper view of the composite is to consider it an assemblage of representative volume elements in the form of a chain of bundles. Each representative volume element may be regarded as an impregnated bundle of fibers. This bundle

has a certain orientation and a certain volume fraction so that all of the bundles differ and must be characterized statistically. Materials models of this type have been analyzed.

COMPRESSION FATIGUE

This discussion has treated static strength, but life prediction also must be considered. Possible approaches follow from an understanding of the dispersed nature of the damage that occurs in composites. Damage takes place at many locations, and damage growth takes place from many locations.

Data demonstrate that the slope of the S-N curve in compression is greatly reduced compared to tension fatigue (Figure 4.2). The greater sensitivity of a unidirectional composite to local imperfections may well be the basis for this result.

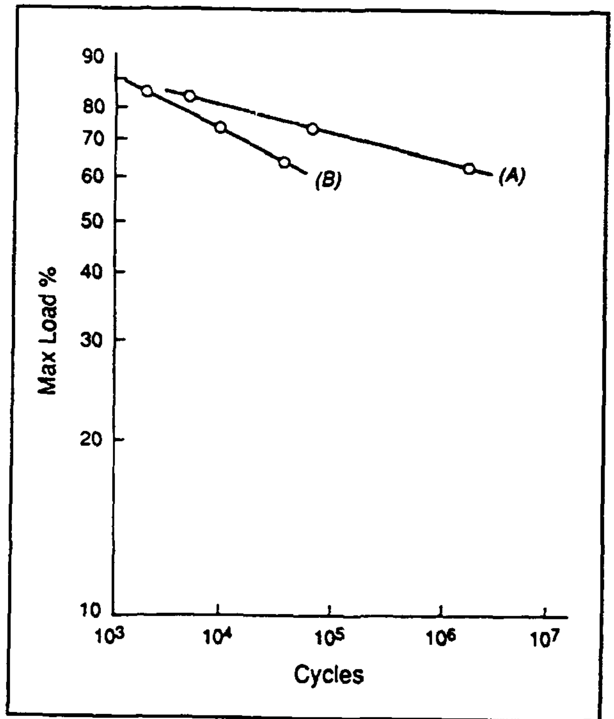

Figure 4.2 Tension fatigue (A) versus acceleration of damage accumulation under compression loading (B) for quasi-isotropic graphite epoxy.

5

STRUCTURAL MATERIALS AND ELEMENTS

The state of the art of life prediction methods for structural materials and elements will be reviewed here primarily at the laminate level of observation and prediction. A great deal of research in this field has been conducted over the past 10 years, and some modeling approaches have been postulated. While these approaches are embryonic in many regards, they do reflect an emerging consensus about an approach to life prediction that may provide engineering results. This chapter is organized in a manner that reflects the committee's perception of that consensus. In particular, the state of the art is presented in the context of the following two premises:

- Life prediction must be based on a clear and precise definition of damage accumulation mechanisms and failure modes as determined from experimental observations.

- Life prediction modeling should be concerned with representative volumes or units of material that control the ultimate failure process, defined by such failure modes.

The literature suggests that there is a relatively small number of distinct damage accumulation mechanisms and failure modes for composite laminates that are in current use (Reifsnider, 1986). This chapter will discuss the damage accumulation mechanisms and failure modes that are thought to be representative of the state of the art and to be well identified and defined at this point. In general, these damage accumulation and failure modes tend to fit under the following three categories:

- Fiber-controlled failures.
- Matrix and interface- (or interphase-) controlled failures.
- Micro- and macroinstability failures.

While the choice of a modeling unit based on observations of damage accumulation mechanisms and failure modes is logically sound and appears to be straightforward, it involves a great deal of technical effort to determine the size, dimensionality, level of observation, failure mechanism or process, and appropriate mechanics analysis to approach the description of the modeling unit throughout the life of the coupon or structural element. These choices are likely to be different for each distinct failure mode, and the degree to which those choices influence the modeling results may depend upon the nature of the material system and the physical loading involved. In addition, the statistical nature of composite materials at the macro and microlevels plays an unusually large role in the choice of analytical modeling procedures and in the results of those efforts. Although a discussion of statistical behavior is usually conducted at the engineering structure level, there is an emerging consensus that statistical variations at the microlevel may be reflected in the characteristic dimension that appears in so many notched strength models that are now widely accepted in this field.

Statistical considerations may also enter the determination of the representative volume chosen for a given failure mode. In addition, the statistical "signature" of any damage or life metric (such as S-N data or property changes) is generally characteristic of the physical process involved and contains essential information for the interpretation of experimental data as well as modeling predictions. Finally, it should be noted that the use of strength or remaining strength as the basis for life prediction is also emerging as a predominant approach. That approach is assumed for most of the discussion in this chapter and is the primary motivation for the selection of failure modes and the representative units defined by those modes as the basis for the discussion.

FAILURE MODES

Matrix Cracking

Matrix cracking is the predominant damage mode for most of the composite material systems currently used (e.g., polymer-, metal-, and ceramic-based materials). Efforts to improve life prediction methodologies should consider the possibility of improved matrix materials demonstrating little or no matrix cracking. Matrix cracking may be a failure mode that defines the life of an element or component in certain circumstances, such as in pressure vessel applications wherein matrix cracking causes leakage failure of the component. However, matrix cracking is more often an initial damage accumulation mechanism that changes the local stress state and induces further damage of other types that defines the terminal failure process and the life of a component. As a failure mode, matrix cracking is largely an initiation problem that is difficult to approach in a generic fashion. Most materials, especially most inhomogeneous materials, contain cracks at some level of observation in the interior or near the surface of the material. The density of matrix cracks in composite laminates often reaches a saturation level, sometimes called a characteristic damage state for matrix cracking (Reifsnider and Highsmith, 1981a). The saturation of crack density in a laminate is demonstrated in Figure 5.1. A radiograph of such a characteristic damage state for a [0,90]s laminate is shown in Figure 5.2. That saturation density (as well as the threshold level of stress or strain at which cracking begins) is defined by the properties of the plies, their thicknesses, and their stacking sequences (Reifsnider and Highsmith, 1981a). To decide on the dimensionality and size of the material unit to be used for modeling, it is necessary to define the size of the matrix crack that defines failure for the application being considered. In some cases, such a crack may be determined by methods of observation; in other cases, cracks may be determined by their consequences, such as leakage of a pressure vessel. The dimensions of the failure crack should be used to determine the dimensions of the unit used for modeling analyses. If cracks with final dimensions initiate as a single physical event, such things as initiation kinetics must be considered. If these final dimensions come from crack growth from latent flaws or initial defects, crack growth considerations must be used for modeling the representative unit. Many such growth concepts have been introduced (Dvorak and Laws, 1986).

An alternative approach to the measurement and growth prediction of individual cracks (which may not be feasible) is a time-averaged approach. Growth data and models are interpreted as a collective effect of multiple crack (or defect) growth and accumulation. For composite materials, the question of instability of matrix cracks is a difficult one. It is typical for matrix cracks to initiate and grow and to be subsequently arrested by a materials phase boundary or other nonuniformity in the material structure. Here again, the final failure event must be used to define the level of instability of interest to modeling efforts. Stiffness changes associated with matrix cracking are also minor on an individual scale, but may be significant when high-density matrix cracking occurs. In this context, it is widely known that characteristic damage states form

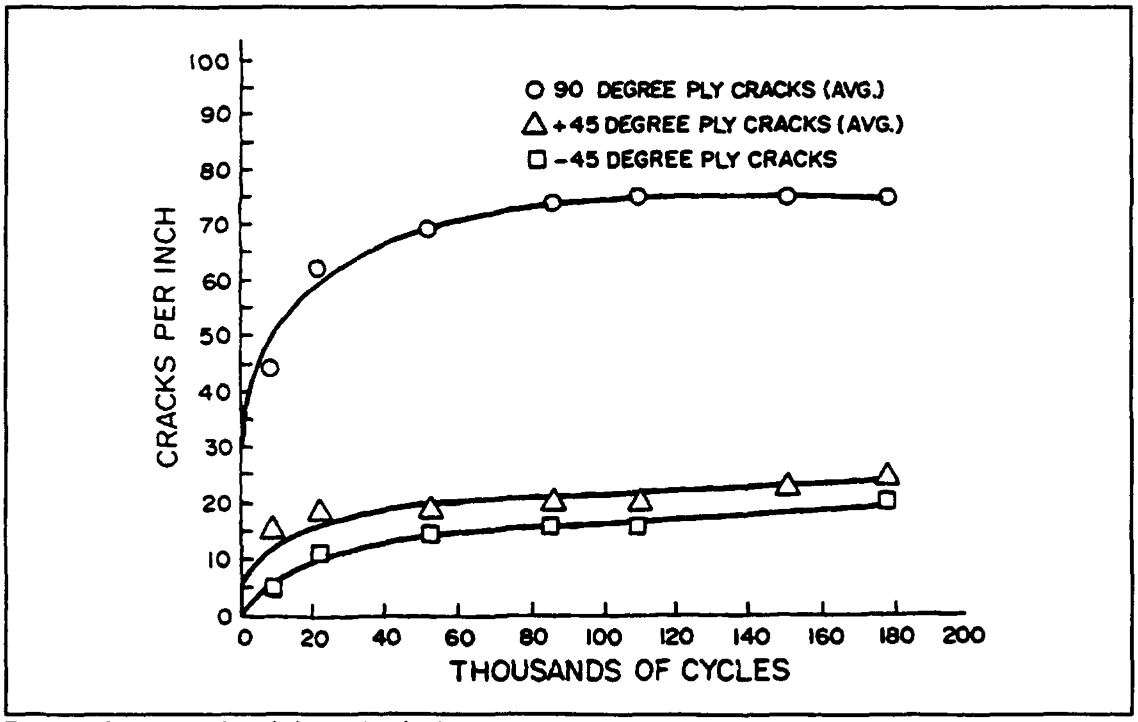

Figure 5.1 Saturation of crack density in a laminate.

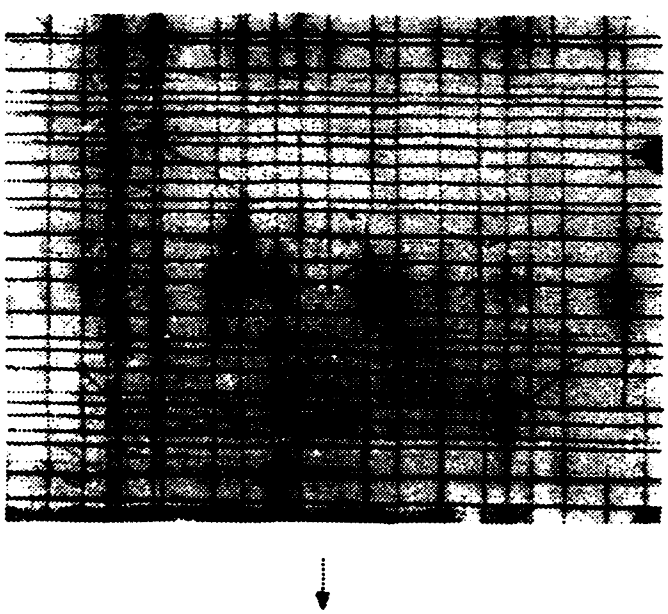

Figure 5.2 Radiograph of saturation cracking in a [0, 90]s laminate.

in laminated composite materials under cyclic loading that includes tensile load excursions transverse to the fiber direction in the plies of the laminate. These characteristic damage states consist of essentially regular arrays of matrix cracks that reach an equilibrium (or saturation) spacing defined by the mechanics of the local situation and the material strength. Stiffness changes can be accurately determined from knowledge of these characteristic states, as can the local stress fields associated with matrix cracks under these conditions (Reifsnider and Highsmith, 1981b).

Delamination

Delamination is perhaps the most widely studied and modeled damage accumulation mechanism and potential failure mode in composite laminates. The literature on this subject is extensive, if not definitive. Delamination most often initiates from latent flaws, initial defects, or matrix cracks. Most treatments of delamination as a failure mode consider only the growth rate of the delamination, assuming that the initial flaws are present. Delamination growth is usually considered to be self-similar crack growth in the fracture mechanics sense. However, a great many additional features associated with the inhomogeneity and anisotropy of composite materials enter this problem. Even if the growth process is predominantly elastic, determination of the correct local stress states is important and may be difficult. The presence of matrix cracks or other damage may significantly alter the stresses and consequent growth rates. In point of fact, delamination growth is rarely completely elastic and rarely consists of planar growth of cracks through matrix material uninfluenced by fibers. Indeed, matrix-dominated growth modes like delamination may be strongly influenced by fiber fracture. It has been established, for example, that delamination growth can be accompanied by extensive local fiber failure, possibly caused by the fracture of fibers that may bridge the delamination. A great many nonlinear effects enter this problem at the practical level. However, representations of stiffness changes associated with delamination and fracture mechanics growth processes (with linear, nonlinear, and even dissipative treatments) are available.

From the standpoint of life prediction, if delamination causes final failure, it may be necessary to consider remaining strength in the presence of delamination. This part of the problem is not as well developed as it might be. For thick laminates and/or large structures, the growth of a delamination may be inconsequential to structural strength (of course, the growth of a delamination to a certain length may exceed arbitrary limits for a given application without consideration of strength). Delaminations that nucleate at edges and cutouts may grow to a finite length and stop growing because they escape stress concentrations or encounter barriers, such as rivets and reinforcements. In other cases, such as shear-driven delamination, single-load-path bonded joints, or narrow coupon specimens, for example, delamination may become unstable and cause structural failure of the component. In any case, most of the analysis tools appear to be available.

Tensile Fiber Failure

In a number of important applications, continuous-fiber-reinforced composite materials fail in a mode that is controlled by tensile fiber failure. This is especially true of applications such as pressure vessels and rotating equipment. Indeed, one objective of the designer is to use these materials in such a way that the superior fiber direction properties can be fully exploited as the controlling elements. This is possible in a number of important applications. From the life prediction standpoint, many aspects of this problem are well in hand, but several remaining factors are very poorly described and extremely difficult to approach. Tensile fibrous composite strength concepts are well developed and comparatively sophisticated. This problem is widely discussed for static loading, and a consensus regarding the dimensions of a proper modeling unit

has clearly emerged (Rosen, 1964; Batdorf and Ghaffarian, 1981, 1983; Bolotin, 1982; Harlow and Phoenix, 1982b; Tamuzs, 1982). Depending on the morphology and properties of the material system involved, those dimensions appear to be no less than a few fiber diameters and no greater than the local damage (or "process") zone around a small group of fiber fractures.

From the life prediction standpoint, the difficulty arises in determining the local stress state and material state after damage has entered the picture. Local stress redistribution can be significant and may have a substantial effect on the remaining strength of the fibers or fiber bundles. This local stress redistribution can be a major part of the problem and may be a substantial part of the life prediction modeling process (Rosen, 1964; Tirosh, 1973; Reifsnider and Highsmith, 1981b; Highsmith, 1984; Kuo and Wang, 1985; Reifsnider and Bakis, 1986).

In particular, the interactions of local stress concentrations associated with matrix cracks, delaminations, and fiber-matrix debonding may greatly influence the calculation of the remaining strength. In addition, fiber bundle strength may be altered by material degradation such as oxidation or other chemical or thermochemical processes.

Microbuckling

Another failure mode that is typical of compression applications is microbuckling. This failure mode consists of stability at the micro or local level. It is usually associated with the formation of kinks in fibers, but may be defined by microbuckling of fiber bundles or plies of a laminate. While this failure mode is widely regarded as a local instability problem under practical circumstances, such things as initial curvature of fibers or plies, and especially of bundles in a woven material, and other imperfections in geometry or spatial variations in material properties may render the local problem less of a *stability effect* and more of a strength effect. This is an important distinction, since the former is more clearly controlled by stiffness and the latter more clearly by material condition or strength. In general, the modeling unit associated with this failure mode is quite local, consisting of dimensions that compare to the kinked region. The process that defines the development of this failure mode usually involves the reduction of lateral support for the microbuckling element (i.e., the development of microcracks in matrix materials, local time-dependent deformations, etc). In general, changes in local stress state tend to be the process that brings about this failure mode.

Global Instability

"Global" is to be interpreted here as at a scale compared to laminate thickness or other laminate dimensions. There are three general subjects within this classification:

- Global stability of defects (e.g., macrocracks under tensile loading).
- Stability in the classical buckling sense.
- Macrocrack stability, which is a well-developed and widely discussed topic with ample literature for the applications community.

There appears to be a limited number of applied situations in which this type of situation determines the life of a component. However, for predominately tensile loading of very thick laminates made from notch-sensitive material serving in the presence of high-stress concentrations, it is possible for a predominant macrodefect to form and grow until stability controls final failure. For those conditions where the size of a cracklike flaw is large compared to the microstructural dimensions of the composite (e.g., to the ply thickness), continuum fracture mechanics techniques are applicable. With the further assumption of small-scale inelasticity at the tip of the flaw, linear elastic fracture mechanics techniques consisting of the strain energy release

rate (or, equivalently in this case, the J integral) or the stress intensity factor can be used with success to represent the phenomenon of flaw growth or at least to estimate flaw instability.

The subject of buckling is also widely discussed and very well developed, and is one of the most sophisticated analytical areas associated with failure. However, the complexity of damage development and other nonlinear effects has not been fully integrated into this field. Indeed, there has been little discussion of the effect of the reduction of stiffness on the subsequent stability of structures at the global level. Considering the number of stiffness-critical designs in which composites typically are applied, this deficiency is surprising.

The unit of material to be considered for the global instability types discussed above differs greatly, as one would expect. In the case of the instability of dominant defects, a local material unit must be considered; for global stability considerations, the geometry and properties of a large structural element are usually modeled. These selections are well documented and thoroughly discussed in the literature.

COMMON FEATURES OF LIFE PREDICTION METHODOLOGY

Life prediction methodology is formulated based on the details of the failure modes discussed above (or others as appropriate). The formulations involve various disciplines, scales of consideration (associated with the unit of material to be modeled), and procedures as necessary. However, there are three generic aspects to the models and modeling approaches. First, the appropriate state of stress must be determined. The geometric level at which this state of stress is analyzed is determined by the scale of the unit of material discussed in the failure mode descriptions mentioned above. In some cases, such as *tensile* failure of fiber bundles, micromechanics considerations may be appropriate. In other cases, such as macrobuckling, the proper boundary value problem may involve the total dimensions of the engineering component. There are at least two dangers to be avoided in this selection process. The first is that unnecessary complexity will be added to the problem by improper choice of the material unit to be modeled, so that the problem becomes intractable. It is not appropriate, in general, to attempt to model the geometry and interactions of all possible matrix cracks in a laminate on an individual and combined basis, for example. However, it may be appropriate to take a representative matrix crack or group of cracks or an assemblage of matrix cracks, delamination, and fiber failures that are representative of the total fracture process for analysis. The second danger is to set the problem at a level that ignores the mechanical and physical mechanisms that are responsible for the final failure process and thereby to construct an analysis that is insensitive to the parameters and details that control the damage accumulation and failure process and ultimately the life of the component being described. From the standpoint of the current state of the art, this may be the most common mistake in attempting to establish a life prediction methodology.

The second general feature of a generic modeling approach is determination of the state of the material. Again, it is essential to choose the material element or unit correctly, as discussed above. The evolution of properties as a function of service life or loading history must be determined under the conditions of interest. This may require considerable modeling effort if mechanistic representations are to be established, so that rational extensions can be made without extensive laboratory testing and evaluation.

The final general feature of life prediction models concerns margins. For a given application and with full knowledge of the statistics of property and performance distributions for a given material system, it is necessary to establish a life prediction methodology that is sensitive to the accuracy and confidence required for engineering calculations in the context of material

variabilities. This is a demanding task, not only in the analytical sense, but especially in the experimental sense since extensive data are required to establish the limits on material variability.

CURRENT PRACTICES

Damage Characterization

The damage accumulation mechanisms and life prediction approach have been distilled from a large body of experimental data on small coupons and elements. The damage accumulation process has been studied using a combination of techniques, with the most useful data being provided by penetrant-enhanced x-ray radiography (Sendeckyj, 1983; Rummel et al., 1980) and deplying (Freeman, 1981). As a result, a consistent picture of the damage accumulation process is now emerging. Nevertheless, these techniques are not being used universally. Considerable reliance is still being placed on nondestructive inspection (NDI) techniques that do not have the necessary resolution to provide useful data. The typical justification for this is that NDI tools which can be used for field and manufacturing inspections will provide data that will be used by industry, and the damage documentation data should be consistent with this. This has led to some amazing statements, such as "impact damage does not grow under severe cyclic loading" and "fatigue failure is a sudden death phenomenon." What is actually happening is that the improper damage documentation tools are not providing information about the actual damage growth.

The situation is further complicated by local phenomena that occur at free edges. In narrow specimens, the edge damage propagates rapidly across the specimen width, and this is often misinterpreted as the true situation for a composite structure. It has been well documented not to be the case. The edge phenomena are restricted to the narrow region near the edges of the specimen. This specimen width effect has been disregarded by many researchers.

Finally, most damage characterization efforts have been conducted using constant-amplitude loading and, without exception, all government agencies that certify high-performance structure demand simulated service testing that is based on random-amplitude testing. It is possible that a random-load history will develop damage states distinct from damage induced by constant-amplitude testing.

Initiation Models

Initiation and accumulation of matrix cracking in model composite laminates have been studied extensively. These studies have shown that the strain at initiation of matrix cracking is strongly influenced by ply thickness and toughness of the matrix material under both static and cyclic loading. The number of matrix cracks increases monotonically with increasing load or load cycles until a characteristic damage state develops. Final failure may be governed by the strength of the zero degree plies or by ply debonding. This process has been successfully modeled using linear fracture mechanics (Chou et al., 1982; Wang et al., 1984, 1985) and simplified mechanics models (Bailey et al., 1979; Highsmith and Reifsnider, 1982, 1986; Dvorak et al., 1985). The models represent the dispersed nature of the matrix cracks. The extension of these models to laminates with practical stacking sequences is complex, and little progress is being made in this direction. These results have a major implication on the design of structural laminates; namely, the threshold strain at which saturation cracking occurs is inversely proportional to ply thickness. For laminates typical of current designs, a ply thickness consistent with desired strain allowables is below desired thickness for fabrication. At a minimum, like plies should not be grouped together if the design is based on a first ply failure consideration. Alternative thinner plies should be considered for very high performance applications.

Delamination Modeling

It is well established that delaminations initiate at stress concentrations (such as free edges, matrix cracks, and impact damage sites) and that their growth can be modeled successfully using classical self-similar crack growth concepts (Wilkins, 1981, 1983b; Russell, 1982; Wilkins et al., 1982; O'Brien et al., 1985, 1988; Martin and Murri, 1988). In brittle matrix composites, the delamination growth rate has a much higher exponent than the crack growth rate in metals. It is also much higher than in tough matrix composites. Effective growth rate is shifted downward without an exponent change by absorbed moisture, test temperature, porosity, and intraply matrix damage for prudently chosen materials.

As a result, even though the mode is more fatigue resistant than monolithic metals, the current design practice for resin matrix composites is to avoid the occurrence of delamination growth by proper static design (including the restriction of design allowables). In situations where this is not done, an otherwise highly fatigue resistant design can be life limited. Problems with delamination normally occur when major out-of-plane loads are missed or improperly treated during the static design. While successful for brittle matrix composites, this design approach has not been validated for tough matrix composites, and slow delamination growth is of current concern. Current emphasis is on avoiding the delamination growth problem by using out-of-plane reinforcement (such as stitching, stapling, and braiding).

Simulation Models

There have been a number of successful simulations of the damage accumulation process in structural laminates. They require the analysis of representative damage states occurring during the damage accumulation process, within the representative volume that is defined by the failure mode to be modeled. This has been done using the finite element, finite difference, variational, and shear lag methods.

Two basic damage growth simulation approaches have been used. In the first, the plies are assumed to have a distribution of latent cracks parallel to the reinforcing fibers (Chou et al., 1982; Wang et al., 1984, 1985). The formation of macromatrix cracks is modeled using the Monte Carlo method in conjunction with stress analyses of matrix crack growth. The initiation and accumulation of matrix cracks in model laminates with different thickness have been successfully modeled using this approach. The model has been extended to include the interaction of perpendicular matrix cracks and the initiation and growth of delaminations from matrix cracks. Final failure of the model laminates has not been predicted to date.

In the second approach, representative damage states are analyzed and used as input to a critical element fatigue model. The model is based on the following assumptions:

- Fatigue behavior of the laminate is governed by the fatigue behavior of a critical element. Fatigue of the critical element is governed by a residual strength degradation fatigue model of the type discussed in the next section.

- Matrix cracking and delamination affect the loading on the critical element.

This model has been used successfully to model the fatigue life and residual strength of structural laminates (Reifsnider and Highsmith, 1981a; Reifsnider, 1982, 1986; Liechti et al., 1982; Miller et al., 1984, 1985; Reifsnider and Stinchcomb, 1986; Stinchcomb and Reifsnider, 1986). Alternate representations of the model geometry may be desirable. The representative volume elements may be portions of the plies, portions of the laminate, or portions of sublaminates, etc.

Such an approach allows, in concept, the use of stochastic variables to define local material properties and microstructural characteristics. However, such studies have not been performed.

Residual Strength and Stiffness Degradation

The residual strength degradation fatigue models (Halpin et al., 1972; Wolff and Lemon, 1972; Jerina and Johnson, 1973; Hahn and Kim, 1975; Yang, 1977, 1978; Yang and Liu, 1977; Yang and Jones, 1978, 1980a, 1980b, 1981, 1982; Yang and Sun, 1980; Yang et al., 1980; Sendeckyj, 1981, in press; Yang and Cole, 1982; Whitney, 1983) currently used are based on the following two hypotheses:

- Instantaneous residual strength is related to cyclic loading by a deterministic equation (defining a residual strength degradation damage metric). The form of this equation is taken to simplify its integration and is consistent with the shape of both constant-amplitude and random-amplitude S-N curves.

- The initial static strength is Weibull distributed. The assumption is made to simplify the derivations.

Based on these two hypotheses, statistically rigorous fatigue models that do a reasonable job of fitting S-N data have been derived. Because of the assumed damage metric, these models can be used to predict fatigue life under two-stage and spectrum fatigue loading by integrating cycle by cycle the governing damage metric equation. Recent statistically rigorous testing of this class of fatigue models has shown that they underpredict the scatter in the fatigue lives and residual strength (Sendeckyj, in press). Recently, Yang et al. (in press) proposed that the parameters in the residual strength degradation equation should be random variables along the lines used in stochastic fatigue growth for metals. They developed such a fatigue model for composites. Based on limited theory-experimental comparisons, the stochastic residual strength degradation fatigue theory seems to properly correlate the scatter in the fatigue lives and residual strength. The aforementioned residual strength degradation fatigue models have the drawback of requiring extensive experimental characterization of each composite laminate considered for a particular application. In an attempt to overcome this drawback, various generalizations of these fatigue models have been proposed. These include the critical element fatigue model, discussed in the previous section, and a stiffness degradation fatigue model. The latter is based on the following hypotheses:

- The laminate contains a critical element the failure of which causes laminate failure. For fiber-dominated laminates under tension-tension fatigue loading, the critical element is the 0° lamina.

- The remaining laminae in the laminate are not critical since their failure does not precipitate laminate failure, but rather influences the loading on the critical element. Their effect is modeled by assuming that the fatigue behavior of the laminate is governed by a stiffness degradation damage metric. As the laminate stiffness degrades, the magnitude of the cyclic loads on the critical element increases.

- The fatigue behavior of the critical and noncritical elements is governed by a residual strength degradation fatigue model.

This class of fatigue models reduces the required amount of experimental characterization. Only the fatigue behavior of the 0°, 90°, and ±45° sublaminates is required for predicting the fatigue behavior of all the laminates containing combinations of these sublaminates. This type of

fatigue model has been tested using an assumed (Sendeckyj, in press) and experimentally measured (Rotem and Nelson, in press) stiffness degradation rate, and it has been shown to correlate experimentally observed trends. The stiffness-degradation-based fatigue models have not been thoroughly validated to date. Nevertheless, they are very promising and imply the possibility of nondestructively determining incipient fatigue failure (Rotem, 1989; Rotem and Nelson, in press).

Continuum Damage Models

Kachanov's continuum damage theory (Kachanov, 1986) has recently been extended to modeling the behavior of composite laminates under static and fatigue loading (Talreja, 1985; Allen et al., 1987a, 1987b, 1987c, 1988; Harris et al., 1988, 1989). In these extensions, the equations of state and constitutive relationships governing the behavior of the composite laminate are assumed to depend on internal state variables in addition to the independent variables. The internal state variables are used to model the consequences of sublaminate damage on a continuum level (i.e., the sublaminate damage is spread out locally and modeled by including additional terms in the constitutive relationships and equations of state). Growth of the internal state variables is assumed to be governed by evolution equations (first-order partial differential equations defining the rate of change of the internal state variables). Based on limited published results, formulation of continuum damage fatigue theories for composite laminates looks very promising.

Cumulative Damage Theories

Three basic types of cumulative damage theories have been proposed for predicting two-stage and spectrum fatigue behavior of composite materials. The first type consists of extensions of the Miner-Palgrem cumulative damage theory for metals fatigue (Miner, 1945; Hofer and Olsen, 1967; Fehrle et al., 1972; Hashin and Rotem, 1978). The second type is derived by cycle-by-cycle integration of the damage metrics in the residual strength degradation-based fatigue theories and their various extensions (Yang et al., 1980; Yang and Jones, 1980a, 1980b, 1981; Sendeckyj, in press). These theories are internally consistent with the underlying fatigue models.

The last type of cumulative damage theory is based on careful experimental observation and simulation of sublaminate damage accumulation (Chou et al., 1982; Wang et al., 1984, 1985). These theories provide good models of damage accumulation during amplitude, two-stage, and even spectrum fatigue loading. Their only weakness is the lack of a laminate failure criterion for tension-tension fatigue loading. For tension-compression and compression-compression fatigue loading, laminate or sublaminate buckling can be used as a failure criterion. As a result, they cannot predict final failure of the laminate subjected to random cyclic loading. Without adequate cumulative damage models, the direct applicability of constant-amplitude laboratory data to structure is in doubt. However, random loading processes have been shown to map in a log linear root-mean-square stress-endurance sense, yielding an expedient characterization technique for today and a strong hint that the mechanistically based damage models could produce usable results.

Applications

The current practice is to apply the residual-strength-based fatigue models to set the spectrum load levels in structural life verification testing and not to perform fatigue design. The shape of the spectrally driven S-N curve for critical structural elements is experimentally characterized using a magnified design fatigue loading spectra. The experimental data are analyzed using a residual strength degradation fatigue model to determine a B-basis S-N curve.

This S-N curve is then used to set either the fatigue spectrum load magnification factor to give a reasonable expected fatigue life for the full-scale test article or a minimum static load that the full-scale test article must survive. If a fatigue test is performed, the full-scale test article is tested using the magnified fatigue spectrum for a predetermined time, usually one or two simulated lifetimes. If the full-scale test article survives this fatigue test, it is concluded that the full-scale structure has been properly designed for fatigue. This approach is fraught with danger, since the more severe spectrum loading may cause fatigue failure of metallic components of the structure. In most cases, the fact that the full-scale test article exceeds the minimum static strength is used as additional evidence that the full-scale structure has been properly designed for fatigue.

SUMMARY

The current state of the art for structural materials and elements defines several issues that require special attention. Current engineering practice at the laminate-element level consists of the experimental S-N characterization for elements such as coupons to establish life versus cyclic load level relationships that are used to avoid fatigue-related degradation and failure. Since composite materials are, in general, so resistant to fatigue degradation, many successful designs have been created with this approach. This policy of avoidance does not address the physical phenomenon of fatigue, however, and cannot address the opportunities for efficiency improvement offered by an understanding of the physical fatigue behavior.

> *Issue:* There does not presently exist any widely accepted approach to lifetime (or residual property) prediction at the structural material or element level. This is a current research area. Mechanistic models and data are needed that include proper representations of service environment effects such as multidimensional stress states, temperature, loading rate, scale, and other factors required by the representative volume element and the corresponding physical process defined by the appropriate failure mode.

Many experimental techniques are available for the detection and characterization of the details of the physical processes associated with fatigue damage development in composite elements. Penetrant-enhanced stereo x-ray radiography, deply methods coupled with scanning microscopes of various types, replications, thermal methods, and other techniques are available. Many of the processes have been investigated, and damage and failure modes have been characterized. The complexity of composite materials and elements causes different damage and failure modes to control residual strength and remaining life under different applied conditions. Hence, it is essential to base characterization and modeling on a sound knowledge and understanding of the operable modes of damage and failure. For example, small (especially narrow and thin) laboratory specimens may be dominated by edge delamination that is unrelated to structural failure modes.

> *Issue:* Much current modeling and testing practice is not generally based on specific knowledge and representation of the physical process or processes that cause property loss and limit life. State-of-the-art experimental methods for the detection and characterization of damage and failure modes such as stereo x-ray radiography, CT scan methods, and thermal techniques are often not available to the applied community or are not used. In many cases, firmly established interpretation methods for newly developed experimental methods are not available. In most cases, the precise details of the microprocesses that control life

are incompletely or erroneously established. Interactions between different damage and failure modes are rarely considered. However, life prediction (or remaining strength prediction) methods that are not firmly based on the analysis and behavior of representative volume elements properly defined by failure modes are severely limited in their generality and ability.

Current practice uses statistical methods to characterize experimental data and to model fatigue behavior, as discussed earlier. The use of statistics in association with life prediction and characterization is extensive and essential. Allowables for static design are based on firmly established statistical methods. Statistical interpretation of fatigue data, at the phenomological level, is widely discussed. However, needs and opportunities also exist in that technical area.

Issue: The statistical nature of fatigue processes and the response of composite materials are not properly or completely represented. Relationships between the statistical behavior of constituent materials and element behavior have not been firmly established. Current model validation methods are generally not firmly grounded on statistical principles. Statistically based fatigue life models are generally not sensitive to details of fatigue damage and failure modes; such models must be calibrated with extensive (and costly) experimental data for all cases of interest. It would be preferable to incorporate statistical representations into mechanistic models that can be extended and extrapolated to unfamiliar circumstances.

Very little life prediction as such is currently attempted in the structural design community. Predicting crack length as a function of cycles of loading is often not appropriate for composites in which many damage events accumulate to control remaining strength and life. In those cases, such as delamination growth, where self-similar crack growth life prediction methods may be appropriate, the slopes of the growth rate curves are generally so high that the accuracy of predictions based on laboratory characterizations is often nebulous. Consequently, current life prediction methods often revert to avoidance procedures, as mentioned earlier. In the technical sense, it is difficult to define life, since many different physical processes may be involved. It is more rigorous to define properties such as strength or stiffness than to attempt to define a unique material or component life characteristic in terms of the number of cycles (or time) to failure in the classical sense used in such typical methods as Miner's rule. In fact, life can be defined by the coincidence of remaining strength and applied stress or strain, and the remaining strength can be evaluated based on specific knowledge of damage and failure models and analysis, as discussed earlier. This approach appears to offer a very good opportunity for progress in this area.

Issue: Current modeling and philosophy focus on life prediction, which is a poorly defined technical problem. Life is more properly defined by changes in such properties as strength or stiffness. Property changes can be interpreted by performance simulation models that relate such changes to life, reliability, and other performance characteristics.

6

STRUCTURAL RESPONSE AND DESIGN PRACTICE

Aircraft structures can be divided into two categories: fixed wing and rotating component vehicles. Fixed wing designs can be further subdivided into highly loaded structures (e.g., military fighters, military transports, commercial transports) and lightly loaded structures (e.g., private aviation). Pressure vessels are also commonly used as aircraft structural components. These tension-loaded structures are an important special case and are also discussed in this chapter.

Compression failure is the primary mode of concern for structures made of composite materials both for fixed wing aircraft and the nonrotating portions of helicopters. Tension failure modes are of primary concern in helicopter rotary structure because tension loads predominate in safety critical rotating components.

The methodologies used to ensure safety of flight and economic durability depend on the type and magnitude of the structural load condition in which potential failure is likely to occur. Therefore, design practices differ for the three classes of aircraft and for pressure vessels. This chapter considers the structural load environment, the associated structural response, and the different strength and life prediction and verification methodologies used in the three classes of aircraft design.

STRUCTURAL LOADING ENVIRONMENT

Structural components made of composite materials experience a complex loading environment consisting of mechanical loads and chemical and thermal "loads". During service life, the properties of the polymer matrix material may change slightly with time. Therefore, the response of the structure to the load and chemical environment may vary with time. The load environment experienced by structural components in the different classes of aircraft also is considerably different.

Fixed Wing Environment

Structural components designed for fixed wing aircraft are well known to experience a complex loading environment. Besides the normal air pressure and vehicle maneuver loads, the designer must also consider vibration loads. Multiple directional loading is normally in membrane directions, but unforeseen out-of-plane loads often are more severe because polymeric composites are inherently weak in that direction. While potential fatigue failures in tension-loaded surfaces are of concern, the primary problem areas are the compression-loaded surfaces. The maximum number of fatigue loading cycles is generally considered to be in the range of 10^6.

Rotating Component Load Environment

Helicopters are rightfully called "expensive fatigue machines". Significant in-plane and out-of-plane forces occur with every rotation of every blade. These forces are due to both aerodynamic pressures and gyroscopic forces since the rotor does not always rotate in a plane normal to the mast. A primary design task is to attempt to isolate the vibratory forces from the vehicle's crew and passengers. Nodal beams, inertia dampers, and vibrating pendulums successfully protect people; however, they generally do not reduce local loads. The oscillatory stresses resulting from these vibratory forces are large relative to the mean stress. Thus, these forces dominate the fatigue loading spectrum. The number of cycles per hour is also high--above 10^5.

A second area where the rotary wing loading environment differs from the fixed wing loading environment is a result of material application. Rotor and dynamic components are good choices for being manufactured of polymeric composite materials because many attractive and benign, but visible, failure modes occur in composites loaded in tension. However, these components experience large axial loads coupled with high out-of-plane loads. The out-of-plane loads are reacted largely by the matrix. Since theoretical analysis techniques relating interlaminar stress with laminate coupling and free edge stress are just evolving to the practical design level, design practices in this area are largely empirical. Often, the analysis is completed after the failure modes are revealed during component tests.

Pressure Vessel Load Environment

Pressure containment structures include a large variety of structures such as pipes, tubes, and tanks. Their load environment usually is entirely tension load dominated with all primary loads being membrane loads. Pressure vessels must sustain both static loads and cyclic loads, generally much fewer than 10^6 cycles. Environmental operating conditions may be severe. Pipes made of composite materials may contain gases or fluids that can be toxic or corrosive to the liner often used in such pipes.

Combined "Loads"

In the context used here, combined "loading" means the combination of mechanical loading and the thermal and chemical environments to which the vehicle is exposed. Polymer matrix material properties may continue to change slightly during service. In addition, polymer matrix materials, to a greater or lesser extent, change because of the external thermal, chemical, and electromagnetic environments. Polymers change with time because they absorb moisture and ultraviolet radiation and react with chemical solvents. The effects of these often complex changes and interactions must be accounted for when designing composite components to meet desired strength and life.

STRUCTURAL RESPONSE

When considering structural response, a primary difference between composites and metals must be kept in mind. Metals are inherently ductile, which permits yielding at points of stress concentration. This is beneficial during static loading, but repetitive yielding is damaging during fatigue loading. Composite materials are inherently brittle and fail under static loading at points of stress concentration. However, progressive microcracking of the matrix at points of stress concentration occurs during fatigue loading, causing redistribution of the load, reduction of the effective stress concentration, and an increase in residual strength. Matrix failure can grow to a

size that causes failure, but this is not the general case. For this reason, it is generally true that if the composite component passes the static strength requirements, it will exceed its fatigue loading requirements.

Failure Criteria

Failure is traditionally defined as either the inability to sustain load or as an unacceptable loss in stiffness. Metal structures generally fail to sustain increased load prior to significant loss in stiffness. Generally, there is no outward evidence of damage other than buckling prior to failure. While glass composite structures may show significant loss in stiffness prior to rupture, carbon composite structures generally do not. Under fatigue loading, both glass and carbon composite structures exhibit visible damage in terms of surface cracking or matrix crazing and show ultrasonic evidence of internal delamination, all prior to a significant loss in strength.

Analysis of metal structures assumes a damage tolerance philosophy with a crack assumed to exist in the most critical area. The crack size is assumed to be the largest that has a probability of not being detected during manufacturing. The material and operating stress levels must be chosen such that the critical crack length is detectable during inspection prior to becoming critical. This approach encourages selection of materials that demonstrate superior toughness. A similar design philosophy is appropriate for composite materials except that the critical flaw size usually is so large that subcritical flaws are not emotionally tolerated. Delamination, surface crazing, and surface splitting can exceed limits of prudent consideration without significant loss of strength or stiffness of the component.

Testing of the dynamic components of helicopters (i.e., grip, spindle, rotor blade, and hub) dramatically demonstrates this point. The damage tolerance criteria for composite structures must define the prudent limits of damage without the usual constraint of eminent rupture. It is insufficient to establish composite material behavior in terms of stiffness and rupture strength. Fracture toughness and knowledge of the occurrence of microcracking and delamination are required. Composite structures fail under monotonic load as a result of buckling or cracking, delamination, and rupture due to a local stress riser or discontinuity. Composite structures fail in fatigue due to sublaminate buckling, loss of stiffness, or strain concentration in the load-carrying fibers resulting from the cumulative effect of microcracking and delamination. Fatigue-loading-induced damage, in terms of microcracking and delamination, often is evident long before loss of strength or stiffness is significant. This is a most desirable characteristic for aircraft structure because it enables meaningful periodic inspection. However, damage state, defined structurally rather than in terms of crack length and rupture, now must be used to define component life and material performance.

Failure Observables

The relatively small number of unique composite material damage states and set specific methods to identify those damage states in composite structure were reviewed in Chapter 5. The heuristic knowledge of the relationship of damage state to structural performance metrics, such as strength, stiffness, and functional integrity of composite structure has been fundamental to formulation of design approaches. Because of extensive experience with component and full-scale structure test data in the design community, knowledge of many of the more subtle failure processes was acquired before adequate analysis emerged.

The damage states observed in laboratory test, qualification test, and service form a basis for the design practices that have evolved. Evidence of damage, such as noise and stiffness loss, has been observed in test, and although study of these phenomena has resulted in a reasonable

degree of understanding, such data are not easily used for monitoring damage growth in actual structures.

DESIGN APPROACHES

Design approaches for nonrotating, rotating, and pressure vessel components are considerably different. Awareness of these approaches is important for understanding the state of the art of present static strength, stiffness, and life prediction methodologies. For general-aviation aircraft, aerodynamic requirements produce deep structural sections for which the structural load intensities are much lower than encountered in commercial transports and military vehicles. Because of the low structural load intensity, general-aviation aircraft wing and fuselage components tend to be thin inner and outer composite surfaces separated by foam or honeycomb core stiffening. Stringers, longerons, and bulkheads are used as stiffeners and for load transfer. Sometimes, such as in the Voyager, wing structure is quite nonhomogeneous. The Voyager's wing design consists of high-percentage 0°-ply spar caps connected by a + and -45° ply-dominated, shear-carrying skin. In commercial and military vehicles, not only are loading intensity and loading rates much higher, but fuselage and wing structures are usually internally pressurized. Therefore, wing and fuselage components tend to be designed as highly loaded, homogeneous skins although again stiffened by stringers, longerons, and bulkheads.

A combination of design approach errors has resulted in microcracking becoming a significant problem in composite structures. These errors include the zeal of materials engineers to demonstrate higher unidirectional properties through minimum resin content, manufacturer's desire to reduce costs by using thicker plies and grouping similar plies together, and overriding "first-ply failure" during computation of laminate strength. All of these actions contribute to low resistance to microcracking (Schutz and Gerhatz, 1977; Parvizi et al., 1978; Lagace and Nolet, 1986).

Such structures as helicopter or tilt rotor hubs, spindles, and rotor grips as well as the space shuttle's solid rocket motor case possess thick laminates, where layers of material of the same fiber orientation are placed contiguously because of manufacturing economy or machinery and process limitations. These thick layers possess low tolerance to transverse strain, whether due to applied load, temperature change, or shear strain due to local deformation caused by impact. Each of these parts has experienced excessive microcracking, which has forced redesign. The first direction in redesign is to reduce the number of contiguous plies of the same orientation. When this route is exhausted, a layer of adhesive is often added at the interface between the thick layers and the adjacent plies. Although these approaches offer significant improvement, they must be considered a "fix" after design. As such, they are less effective than they might be if adequate analysis methods had existed for their consideration in initial design.

Probably the most challenging problem when designing with a brittle material is loss of structural efficiency due to attachments. The most efficient design is one where major load transfer is accomplished within a single laminate. Shell construction uses shaping of a single laminate to meet both functional and structural objectives. Mechanical joints should be either nonexistent or treated very conservatively. Composites are finding application in the construction industry in large cooling towers and corrosion-resistant or electrically transparent buildings. These applications exploit the ability to mold large complex structures and to separate the bending material from the shear material. Mechanical attachment, when required, is confined within the shear material.

Figure 6.1 illustrates one design solution for the elimination of joints. This is a typical beam member, but is as unique to composites as the "I" beam is to metal. First, it is a single laminate that can be assembled on a simple tool. Second, the same laminate contains both tension and compression beam caps together with the shear web. The concept is already proven whether applied to a stiffener, ring frame, or a wing box. A second design solution is to blend functional features with structural requirements through molding. Figure 6.2 illustrates a section of a fuselage wherein the conventional terms skin, bulkhead, and longeron are mixed with shelf and equipment box, losing their identity together with the strain concentration and weight of their individual attachment requirements. There are general guidelines for designing and manufacturing components of composite materials. A good design is one that delays the onset of microcracking in both strain level and cyclic life rather than "conservatively" accepting poor performance of pseudoisotropic laminates that are homogeneous across their width. The following points summarize one effective design approach. First, design structures with laminates possessing a high percentage of 0° plies for axial or bending loads and separate these members by laminates of all ±45° plies for shear loads. The ±45° laminate offers a degree of damage containment to the 0° members while possessing a high degree of damage tolerance. Second, the 0° plies should not exceed 60 percent, and the angle plies should be interspersed so that plies of the same orientation are never together. Third, fasteners should not be installed in laminates containing a high percentage of 0° plies. Attachment holes should be placed in all ±45° laminates where possible. Fourth, the ability of continuous 0° plies to carry axial load and their need to be straight should be exploited. At intersections of axial members, such as rotor hubs or at the corners of large cutouts, material, either angle plies or cut zero plies, should not be added to form a radius. Corner radii are good practice in metals and bad practice in composites. In composites, strength decreases with increasing radius. The load in cut fibers aligned with the principal load must be distributed to the uncut fibers in the same manner as at a major cutout in a tension field web. Fifth, designs prone to fail in peel should be avoided. The flexure capability of the laminate should be utilized to offer compliance and to mitigate out-of-plane loads. Sixth, the low transverse shear stiffness and strength should be accounted for when considering out-of-plane loads. It should be remembered that time under load is significant because this property is matrix dependent. Impact loads, whether from dropped objects or rapidly tightened fasteners in joints containing gaps, can cause local shear microcracking, whereas a more gentle loading with the same resulting deformation may not. Seventh, laminates should be selected from lamination patterns that do not result in significant strain transverse to the major load directions. Generally, plies perpendicular to major loads are not required and should be avoided.

Composite pressure vessels are generally fabricated using the filament winding technique. They can be lined or unlined. Lined vessels are designed for chemical handling or are used as high-efficiency, high-pressure containers for fluids. Unlined vessels broadly include a variety of pipes, tubes, and tanks. Under fatigue or sustained static loads lower than the ultimate failure pressure, unlined vessels usually have the resin-controlled failure mode of either leaking or "weeping" caused by either fiber-matrix bond failure or matrix microcracking. These vessels usually are not designed for minimum weight, but only to prevent the resin-controlled failure condition. Metal-lined, high-pressure vessels used in military and aerospace applications usually are made with a high-performance fiber in an epoxy matrix. Such vessels experience much higher stress levels than unlined vessels. The mode of failure of lined vessels under long-term loading usually is fiber controlled and catastrophic. This is particularly true for vessels containing high-pressure gases. Catastrophic failure must, of course, be avoided. The design approach for assuring the required service lifetime is mostly based on strength. Typically, a safety factor of two or more is applied to the vessel's static burst pressure. Safety factors are selected based on experience and published lifetime data, when available.

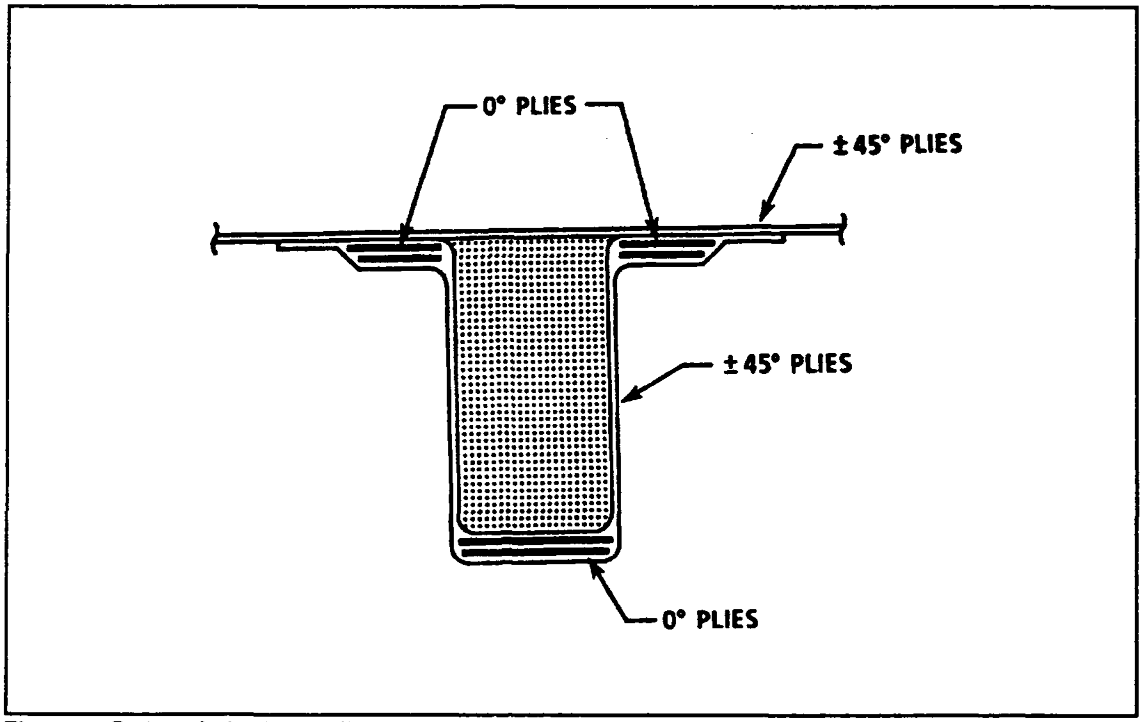

Figure 6.1 Design solution for the elimination of joints in aircraft.

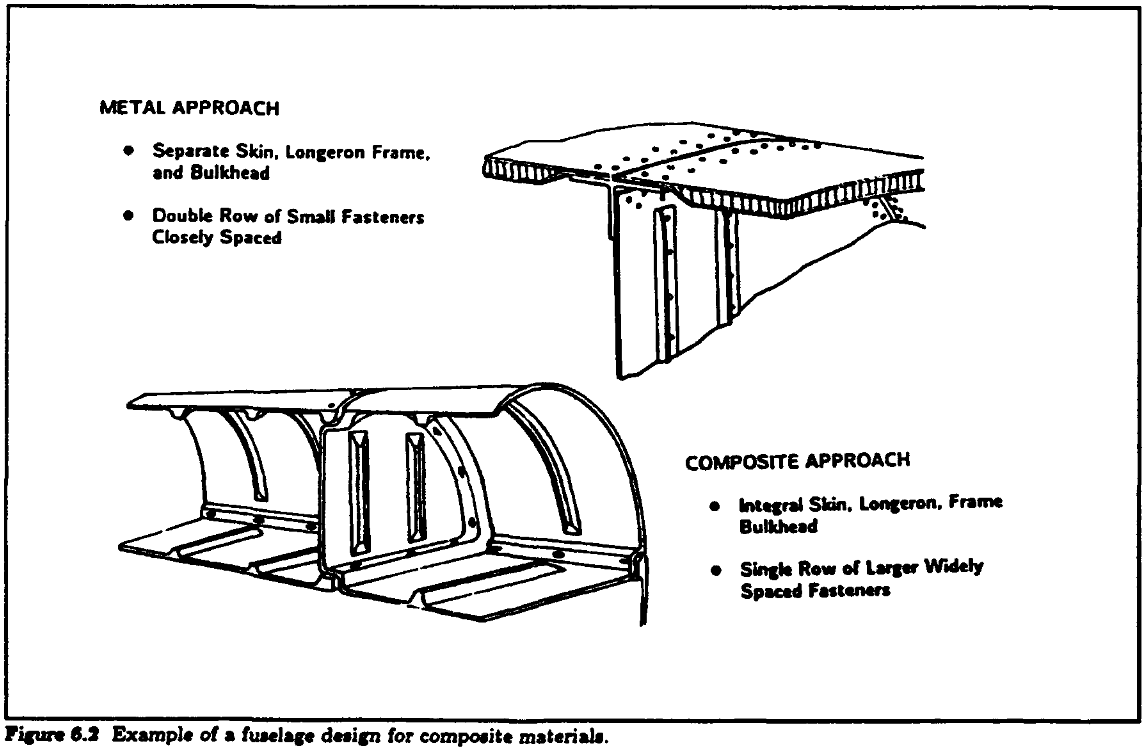

Figure 6.2 Example of a fuselage design for composite materials.

LIFE PREDICTION METHODOLOGY

The aerospace engineering community has evolved practical, although not rigorous, procedures for demonstrating adequate service life for structures made of composite materials (Lauraitis et al., 1981; Horton and Whitehead, 1988; O'Brien, 1988; O'Brien et al., 1988). Currently there are three application-specific approaches: military, civil aviation, and rotorcraft. Although the specifics of each approach differ, they have two principles in common. First, due to uncertainties in the translation of coupon data to components, service life is demonstrated at the component level. Second, the large scatter in fatigue life is addressed by increasing the spectrum load level such that, typically, testing for two lifetimes becomes statistically significant. Damage tolerance criteria that reflect the potential of catastrophic failure in compression constrain most aircraft design. As a result, adequate service life may not be, so far, a critical issue for many structural components made of composites.

The civil aviation industry, under the direction of the Federal Aviation Administration, establishes life in terms of the strain number of cycles for a structure at the subcomponent level. A scatter factor of two with a load amplification factor on the randomized load spectrum is used at the subcomponent level. A sample of 30 specimens is used to establish the mean life that together with two additional samples of five each at different load levels establishes the shape of the life versus spectrum curve for the material and structural concept. These data are used to establish the factor that when applied to the load spectrum demonstrates a structural life to the desired confidence level when the full component is tested to two lifetimes. This same component is tested to an additional two lifetimes with inflicted damage, but without the extra factor on load. The full airframe is subjected to two lifetimes of spectrum loading without the extra load factor, primarily to confirm the life of the noncomposite structure elements in the airframe.

The military aviation industry uses a similar approach at the subcomponent level to locate and correct any fatigue-sensitive design features. Damage tolerance constraints generally preclude use of fatigue-sensitive design concepts. Adequate life is demonstrated at the component level through multiple tests where each component is subjected to two lifetimes of spectrum fatigue without a factor on load and often with inflicted or simulated damage. If the component survives, a second set of two lifetimes of spectrum load is applied, but with a factor on load. If the component survives, the load factor is increased and another two lifetime blocks of loading are applied. This process is repeated until failure is achieved, but failure is generally not in the composite material.

Rotorcraft dynamic components are tension-dominated applications where the lack of damage tolerance constraints results in more aggressive use of composite materials, resulting in benign, but life-limiting failure modes. Simulation of inflicted or initial damage considered is not applicable, partly because of the massive nature of the structure, but also because of the extreme care exercised in practice by rotorcraft personnel on cyclically loaded structures. Adequate life is demonstrated at the component level by testing to two lifetimes of spectrum loading with a load factor to account for environmental effects and scatter. A successful replication of six components is required to qualify the component. Benign failure in terms of matrix microcracking and delamination is the rule; definition of an allowable damage size remains an unresolved issue.

These three engineering certification approaches have been driven by the lack of a definitive life prediction methodology. The need for cumulative damage procedures is obviated by using simulated service load and environmental spectrum testing. The approaches neither directly contribute to the design process nor provide a basis for understanding unexpected

problems that may arise during service. Furthermore, they often result in an expensive evaluation and demonstration process.

Demonstration of adequate service life sometimes is based on damage initiation and growth data using fracture mechanics principles (Wilkins, 1983a, 1983b). This procedure is far less commonly used than approaches based on static strength and experimental verification of fatigue life. The damage growth analysis procedure has been considered to be too uncertain and costly. In addition, the approach is often considered to be impractical because of difficulties in accurately determining flaw growth models for a real damage condition in a structure; uncertainty in the relationship between damage state and incremental damage growth; large slope in the damage state versus incremental growth data results in small slope errors giving large errors in life prediction; and large scatter in fatigue life that occurs at any stress, which can lead to early failure. These concerns can be overcome, however (Wilkins, 1983a, 1983b). Although analytical approaches to damage growth would save considerable cost in structural qualification, existing procedures are not extensively used.

Fixed Wing Aircraft Components

The approach most commonly used to account for fatigue life in fixed wing aircraft is based on static strength analysis. If static strength requirements are met, fatigue life requirements are also assumed to be met. An extensive test and service data base for epoxy-matrix composite materials supports this assumption. The reason for using the static strength approach is that the cost of collecting damage growth data and developing fatigue analysis methods is large while their value remains unproven. These cost considerations are a significant impediment to changing design practices because reduction of vehicle cost is a major goal. They will remain an impediment to change until other approaches are developed and proven to be technical and cost-effective improvements. In this static design approach, no assumption is made that damage will not develop or grow during service. However, damage growth is assumed to arrest, in a benign manner, during the cyclic life. This is equivalent to saying that the spectrum stress-life fatigue curve will show only run outs. Experimental data for brittle epoxy-matrix composites, the materials most commonly used today, support this assumption. Damage arrest usually occurs in these materials because stress and strain levels at the damage boundary decrease as the damage size increases.

The strongest lesson learned for brittle epoxy-matrix composites is that conventional laboratory coupon fatigue data may not directly apply to structures. There are three reasons for this. First, failure modes that occur in a laboratory coupon are different from those that occur in a structural component. Second, the strain levels at which coupon fatigue failure can occur often are considerably higher than the maximum static strain level allowed in the structure. Third, the analyses used for correlation were not based upon rigorous failure mode analysis (i.e., the effects of damage and damage scale on structural response). Thus, coupon fatigue failures are not considered important. This can, and does for many engineers, mean there is no general need for fatigue life prediction analysis. This does not imply that collection of fatigue data is unnecessary. Some limited, and often useful, correlations between coupon data and structural response can be made given adequate failure mode correlation. Because of the first-order macroscopic coupling of defect state with load and geometry, life prediction analysis must commence with the isolation of the failure mode. The existence of a "geometry-independent material property" may be valid only for the fiber tension failure process that has a very small characteristic dimension.

Rotary Components

Fatigue life estimations for composite structures are currently calculated using Miner's linear fatigue damage hypothesis. Life is determined by correlating the calculated loads data with S-N curves in conjunction with frequency-of-occurrence spectrum. When component test data are not available, published fatigue data, coupon test data, or fatigue test data for similar components are used for fatigue life estimation. A Soderberg diagram correction for steady stress is used where the component steady stress is significantly different from the coupon steady stress or similar component steady stress. Where test results show evidence of more than one mode of failure, the fatigue life of the component is calculated on the basis of each failure mode. The lowest life is then considered applicable.

Pressure Vessels

Current life estimation methodology for metal-lined pressure vessels typically is a fairly simple procedure. Experimental verification of life prediction methodology on pressure vessels is expensive. This is because the vessel specimens themselves are expensive, the minimum number of vessels required for a lifetime experiment can easily be over a hundred, and the time involved for testing is in terms of years. Substantial work has been done on an aramid-epoxy and S-glass-epoxy composite both on materials and small spherical and cylindrical vessels at the Lawrence Livermore National Laboratory (Glaser et al., 1983, 1984). For engineering purposes, these data are probably adequate to represent the two composite systems generally. In specific cases, limited testing of a few vessels to spot-check against the published results may be necessary. For graphite composites, much less data are published, even just on the fiber-matrix materials. However, the need for extensive data comparable to the data on aramid-epoxy composites is debatable. This is because graphite fibers are among the most inert and stable materials known. Furthermore, available lifetime data on fiber-epoxy tows show that its life distribution is the broadest among composites, indicating that it will be even more expensive to obtain data of statistical significance than for the aramid and glass fiber composites.

OPPORTUNITIES AND ISSUES

Although not necessarily ineffective, the state of technological development of fatigue life assurance procedures used in the aerospace industry is still somewhat primitive. The procedures are largely empirically based, and there is concern that they are more costly than necessary. There also is the nagging feeling that because of unknown failure modes, the industry may be in trouble and not know it. Opportunities for improving life assurance procedures are summarized in this section.

First, analysis techniques that, by failure process, relate coupon data to full-scale structure need to be further developed. Much of the required analysis exists, but has not been well documented and summarized. Such information would help avoid some of the necessity of testing large structures. The stress distribution at damage regions redistributes as geometry scale increases; this is one reason why coupon data do not relate to structure. Other examples of the correlation problem between coupons and structure are: compression strain to failure after impact of coupons is significantly lower than that for structural elements; fatigue life curves change slope as coupon geometry changes; wide coupons fail at longer fatigue lives than narrow coupons because edge-induced delamination stops growing; fatigue damage, which leads to failure, starts in the center of large unnotched panels, not at the edge as in narrow coupons (Ryder and Walker, 1979; Ryder and Lauraitis, 1981). Because of these known interactions between defect state, specimen scale, and incremental damage growth, simple geometry-independent, cycle-counting

procedures such as those being used in the rotorcraft industry are known to be inadequate. Simplified procedures need to be developed to account for spectrum effects on damage growth at the preliminary design level.

Second, during the design process, damage containment is often built into the structure. To accomplish damage containment efficiently during the design process, simple analytical models are needed that allow prediction of whether damage will grow under static loads, understanding of damage containment mechanisms, and prediction of lamination stacking sequences that give the best stress state.

Third, the time-dependent properties of thermoplastic resin systems need to be studied. The importance of these properties for long-life structures needs to be understood.

Fourth, understanding the stress rupture behavior of the composite material system remains one of the keys to predicting the life of pressure vessels. Using fiber-matrix tow specimens, statistically significant data can be obtained within reasonable cost. If, because the tow specimen contains the entire process zone, stress rupture behavior can be considered a true material property, then it should be independent of structural shape. This means that the large expense required to obtain actual lifetime data on different structural types or sizes can be spared. Life prediction under the same state of loading can thus be adequately based on stress rupture data of the composite material system itself. Available experimental data on S-glass-epoxy and aramid-epoxy systems verify this concept (Glaser et al., 1983). Life estimations based on data from these two composite materials are more conservative for engineering design than that for actual pressure vessels.

REFERENCES AND BIBLIOGRAPHY

Aeronautics and Space Engineering Board. 1987. Advanced Organic Composite Materials for Aircraft Structures--Future Program. Washington, D.C.: Aeronautics and Space Engineering Board.

Allen, D. H., S. E. Groves, and C. E. Harris. 1987a. A Cumulative Damage Model for Continuous Fiber Composite Laminates with Matrix Cracking and Interply Delaminations. Pp. 57-80 in Composite Materials: Testing and Design (8th Conference). Philadelphia: American Society for Testing and Materials. ASTM-STP 972.

Allen, D. H., C. E. Harris, and S. E. Groves. 1987b. A Thermomechanical Constitutive Theory for Elastic Composites with Distributed Damage - Part I: Theoretical Development. International Journal of Solids and Structures. 23(9):1301-1318.

Allen, D. H., C. E. Harris, and S. E. Groves. 1987c. A Thermomechanical Constitutive Theory for Elastic Composites with Distributed Damage - Part II: Application to Matrix Cracking in Laminated Composites. International Journal of Solids and Structures. 23(9):1319-1338.

Allen, D. H., S. E. Groves, C. E. Harris, and R. G. Norvell. 1988. Characteristics of Stiffness Loss in Crossply Laminates with Curved Matrix Cracks. Journal of Composite Materials. 22:71-80.

Allix, O., P. Ladeveze, D. Gilletta, and R. Ohayon. 1989. A Damage Prediction Method for Composite Structures. International Journal for Numerical Methods in Engineering. 27:271-283.

Aronsson, C., and J. Backlund. 1986. Tensile Fracture of Laminates with Cracks. Journal of Composite Materials. 20:287-307.

Aveston, J., A. Kelly, and J. M. Sillwood. 1980. Long Term Strength of Glass Reinforced Plastics in Wet Environments. Pp. 556-568 in Advances in Composite Materials: Proceedings of the Third International Conference on Composite Materials, Volume 1, A. R. Bunsell, C. Bathias, A. Martrenchar, and G. Verchery, eds. Oxford: Pergamon Press.

Backlund, J., and C. Aronsson. 1986. Tensile Fracture of Laminates with Holes. Journal of Composite Materials. 20:259-286.

Bailey, J., P. T. Curtis, and A. Parvizi. 1979. On the Transverse Cracking and Longitudinal Splitting Behavior of Glass and Carbon Fibre Reinforced Epoxy Cross Ply Laminates and the Effect of Poission and Thermally Generated Strain. Proceedings of the Royal Society. 366:599-623.

Bascom, W. D., J. L. Bitner, R. J. Mouton, and A. R. Siebert. 1980. Interlaminar Fracture of Organic-Matrix, Woven Reinforcement Composites. Composites. 11(1):9-18.

Batdorf, S. B., and R. Ghaffarian. 1981. Tensile Strength of Unidirectionally Reinforced Composites. University of California at Los Angeles Technical Report. Report UCLA-ENG-8116(July).

Batdorf, S. B., and R. Ghaffarian. 1983. Size Effect and Strength Variability of Unidirectional Composites. University of California at Los Angeles Technical Report. Report UCLA-ENG-8343(September).

Bauer, R. S. 1986. Toughenable Epoxy Matrix Resins for Advanced Composites. Pp. 510-519 in Proceedings of the 18th International SAMPE Technical Conference.

Beaumont, P. W. R. 1987. The Fatigue Damage Mechanics of Composite Laminates. Pp. 53-63 in Damage Mechanics in Composites, A. S. D. Wang and G. K. Haritos, eds. New York: American Society for Mechanical Engineers.

Bergmann, H. W., and R. Prinz. 1989. Fatigue Life Estimation of Graphite/Epoxy Laminates Under Consideration of Delamination Growth. International Journal for Numerical Methods in Engineering. 27:323-341.

Bolotin, V. V. 1982. Stochastic Models of Fracture of Unidirectional Fiber Composites. Pp. 3-16 in Fracture of Composite Materials, G. C. Sih and V. P. Tamuzs, eds. The Hague: Martinus Nijhoff Publishers.

Bullock, R. E. 1974. Strength Ratios of Composite Materials in Flexure and in Tension. Journal of Composite Materials. 8:200-206.

Chiao, C. C., and T. T. Chiao. 1982. Aramid Fibers and Composites. Pp. 272-317 in Handbook of Composites, B. Lubin, ed. New York: Van Nostrand Reinhold.

Chou, P. C., A. S. D. Wang, and H. Miller. 1982. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-82-4083(September).

Cullen, J. S. 1981. Mode I Delamination of Unidirectional Graphite/Epoxy Composite Under Complex Load Histories. Texas A&M University Mechanics and Materials Center Technical Report. Report MM 3724-81-13(December).

Dvorak, G. J., and N. Laws. 1986. Analysis of First Ply Failure in Composite Laminates. Journal of Engineering Fracture Mechanics. 25(5):763-770.

Dvorak, G. J., N. Laws, and M. Hejazi. 1985. Analysis of Progressive Matrix Cracking in Composite Laminates-I: Properties of a Ply with Cracks. Journal of Composite Materials. 19:216-234.

Eisenmann, J. R. 1976. Bolted Joint Static Strength Model for Composite Materials. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum TM-X-3377.

Eisenmann, J. R., and B. E. Kaminski. 1972. Fracture Control for Composite Structures. Journal of Engineering Fracture Mechanics. 4(4):907-913.

Evans, A. G. 1980. Fatigue in Ceramics. International Journal of Fracture Mechanics. 16(6):485-498.

Fehrle, A. C., J. R. Carroll, and S. M. Freeman. 1972. Development of an Understanding of the Fatigue Phenomena of Bonded and Bolted Joints in Advanced Filamentary Composite Materials. Volume III. Fatigue Analysis and Fatigue Mode Studies. AFFDL Technical Report. AFFDL-TR-72-64.

Freeman, S. M. 1981. Damage Propagation in Graphite-Epoxy by a Deplying Technique. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-81-3157(December).

Freeman, S. M. 1982. Characterization of Lamina and Interlaminar Damage in Graphite-Epoxy Composites by a Deploy Technique. Pp. 50-62 in Composite Materials: Testing and Design (Sixth Conference), I. M. Daniel, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 787.

Glaser, R. E., R. L. Moore, and T. T. Chiao. 1983. Life Estimation of an S-Glass/Epoxy Composite Under Sustained Tension Load. 5(1):21-25.

Glaser, R. E., R. L. Moore, and T. T. Chiao. 1984. Life Estimation of Aramid/Epoxy Composites Under Sustained Tension Load. 5(1):26.

Hahn, H. T., and R. Y. Kim. 1975. Proof Testing of Composite Materials. Journal of Composite Materials. 9:297-311.

Halpin, J. C., T. A. Johnson, and M. E. Waddoups. 1972. Kinetic Fracture Models and Structural Reliability. International Journal of Fracture Mechanics. 8:465-468.

Hancox, N. L. 1981. The Influence of Voids on the Hydrothermal Response of Carbon Fiber Composites. Journal of Materials Science. 16:627.

Harlow, D. G., and S. L. Phoenix. 1982a. The Chain-of-Bundles Probability Model for the Strength of Fibrous Materials--I: Analysis and Conjectures. Journal of Composite Materials. 12:195-214.

Harlow, D. G., and S. L. Phoenix. 1982b. Tight Bounds for the Probability Distribution of the Strength of Composites. Pp. 17-27 in Fracture of Composite Materials, G. C. Sih and V. P. Tamuzs, eds. The Hague: Martinus Nijhoff Publishers.

Harris, C. E., D. H. Allen, and E. W. Nottorf. 1988. Modelling Stiffness Lose in Quasi-Isotropic Laminates Due to Microstructural Damage. Journal of Engineering Materials and Technology. 110(April):128-133.

Harris, C. E., D. H. Allen, and E. W. Nottorf. 1989. Prediction of Poisson's Ratio in Cross-Ply Laminates Containing Matrix Cracks and Delaminations. Journal of Composites Technology and Research. 11(2):53-58.

Harrison, R. P., and M. G. Bader. 1983. Damage Development in CFRP Laminates Under Monotonic and Cyclic Stressing. Fibre Science and Technology. 18:163-180.

Hashin, Z. 1981. Fatigue Failure Criteria for Unidirectional Fiber Composites. Journal of Applied Mechanics. 48:846-852.

Hashin, Z. 1983. A Statistical Cumulative Damage Theory for Fatigue Life Prediction. Journal of Applied Mechanics. 50:571-579.

Hashin, Z. 1985a. Analysis of Cracked Laminates: A Variational Approach. Mechanics of Materials. 4:121-136.

Hashin, Z. 1985b. Cumulative Damage Theory for Composite Materials: Residual Life and Residual Strength Methods. Composites Science and Technology. 23:1-19.

Hashin, Z. 1986. Analysis of Stiffness Reduction of Cracked Cross-Ply Laminates. Journal of Engineering Fracture Mechanics. 25(5):771-778.

Hashin, Z. 1987. Analysis of Orthogonally Cracked Laminates Under Tension. Journal of Applied Mechanics. 54:872-879.

Hashin, Z. 1988. Thermal Expansion Coefficients of Cracked Laminates. Composites Science and Technology. 31:247-260.

Hashin, Z., and A. Rotem. 1973. A Fatigue Failure Criterion for Fiber Reinforced Materials. Journal of Composite Materials. 7:448-465.

Hashin, Z., and A. Rotem. 1978. A Cumulative Damage Theory of Fatigue Failure. Journal of Materials Science and Engineering. 34:147-160.

Herakovich, C. T., and M. W. Hyer. 1986. Damage-Induced Property Changes in Composites Subjected to Cyclic Thermal Loading. Journal of Engineering Fracture Mechanics. 25:779-791.

Hertzberg, R. W., and J. A. Manson. 1980. Fatigue of Engineering Plastics. New York: Academic Press.

Highsmith, A. L. 1984. Damage Induced Stress Redistribution in Composite Laminates. Ph.D. dissertation. Virginia Polytechnic Institute and State University, Blacksburg, Va.

Highsmith, A. L., and K. L. Reifsnider. 1982. Stiffness-Reduction Mechanisms in Composite Laminates. Pp. 103-117 in Damage in Composite Materials, K. L. Reifsnider, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 775.

Highsmith, A. L., and K. L. Reifsnider. 1986. Internal Load Distribution Effects During Fatigue Loading of Composite Laminates. Pp. 233-251 in Composite Materials: Fatigue and Fracture, H. T. Hahn, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 907.

Hofer, K. E., Jr., and E. M. Olsen. 1967. An Investigation of the Fatigue and Creep Properties of Glass Reinforced Plastics for Primary Aircraft Structures. IIT Research Institute Final Report. NOW 65-0425-F(April).

Horton, P. E., and R. Whitehead. 1988. Damage Tolerance of Composites. Volumes I and II. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL TR-87-3030(July).

Hull, D., and P. J. Hogg. 1980. Nucleation and Propagation of Cracks During Strain Corrosion of GRP. Pp. 543-555 in Advances in Composite Materials: Proceedings of the Third International Conference on Composite Materials, Volume 1, A. R. Bunsell, C. Bathias, A. Martrenchar, and G. Verchery, eds. Oxford: Pergamon Press.

Humphreys, E. A., and B. W. Rosen. 1979. Development of a Realistic Stress Analysis for Fatigue Analysis at Notched Composite Laminates. Houston, Tex.: National Aeronautics and Space Administration. NASA CR-159119(May).

Jerina, K. L., and T. A. Johnson. 1973. Characterization of Composites for the Purpose of Reliability Evaluation. Pp. 5-64 in Analysis of Test Methods for High Modulus Fibers and Composites, J. M. Whitney, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 521.

Johnson, W. S., and P. D. Mangalgiri. 1985. Influence of the Resin on Interlaminar Mixed-Mode Fracture. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum 87571(July).

Kachanov, L. M. 1986. Introduction to Continuum Damage Mechanics. The Hague: Martinus Nijhoff Publishers.

Kenney, M. C., J. F. Mandell, and F. J. McGarry. 1985. Fatigue Behaviour of Synthetic Fibres, Yarns, and Ropes. Journal of Materials Science. 20(6):2045-2059.

Kinloch, A. J., and R. J. Young. 1983. Fracture Behavior of Polymers. London: Applied Science Publishers.

Kinloch, A. J., S. J. Shaw, and D. L. Hunston. 1983a. Deformation and Fracture Behaviour of a Rubber-Toughened Epoxy: Failure Criteria. Journal of Polymer Composites. 24(10):1355-1363.

Kinloch, A. J., S. J. Shaw, D. A. Tod, and D. L. Hunston. 1983b. Deformation and Fracture Behaviour of a Rubber-Toughened Epoxy: Microstructure and Fracture Studies. Journal of Polymer Composites. 24(10):1341-1354.

Kramer, E. J. 1979. Environmental Cracking in Polymers. Pp. 55-120 in Developments in Polymer Fracture, E. H. Andrews, ed. London: Applied Science Publishers.

Kunz, S. C., and P. W. R. Beaumont. 1981. Low-Temperature Behaviour of Epoxy-Rubber Particulate Composites. Journal of Materials Science. 16(1):3141-3152.

Kunz-Douglass, S., P. W. R. Beaumont, and M. F. Ashby. 1980. Model for the Toughness of Epoxy-Rubber Particulate Composites. Journal of Materials Science. 15(5):1109-1123.

Kuo, A. Y., and S. S. Wang. 1985. Dynamic Hybrid Finite-Element Analysis for Interfacial Cracks in Composites. Pp. 5-34 in Delamination and Debonding of Materials, W. S. Johnson, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 876.

Lagace, P. A., and S. C. Nolet. 1986. The Effect of Ply Thickness on Longitudinal Splitting and Delamination in Graphite Epoxy Under Compressive Cyclic Load. Pp. 335-360 in Composite Materials: Fatigue and Fracture, H. T. Hahn, ed. Philadelphia: American Society for Testing and Materials. ASTM STP 907.

Lauraitis, K. N., J. T. Ryder, and D. E. Pettit. 1981. Advanced Residual Strength Degradation Rate Modeling for Advanced Composite Structures. Volume II, Tasks II and III. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-79-3095(July).

Law, G. E., and D. J. Wilkins. 1984. Delamination Failure Criteria for Composite Structures. Naval Air Systems Command Technical Report. NAV-GD-0057(May).

Lee, J. W., I. M. Daniel, and G. Yaniv. 1989. Fatigue Life Prediction of Cross-Ply Composite Laminate. Pp. 19-28 in Composite Materials: Fatigue and Fracture, P. A. Lagrace, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 1012.

Leve, H. L. 1969. Cumulative Damage Theories. Pp. 170-203 in Metal Fatigue: Theory and Design, A. F. Madayag, ed. New York: J. Wiley & Sons.

Liechti, K. M., K. L. Reifsnider, W. W. Stinchcomb, and D. A. Ulman. 1982. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-82-4094.

Liu, S., and J. A. Nairn. 1990. Fracture Mechanics Analysis of Composite Microcracking: Experimental Results in Fatigue. In Proceedings of the Fifth Technical Conference on Composite Materials. East Lansing, Mich.: American Society for Composites.

Liu, S., and J. A. Nairn. In press. The Formation and Propagation of Matrix Microcracks in Cross-Ply Laminates During Static Loading. Journal of Composite Materials.

Lorenzo, L., and H. T. Hahn. 1986. Fatigue Failure Mechanisms in Unidirectional Composites. Pp. 210-232 in Composite Materials: Fatigue and Fracture, H. T. Hahn, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 907.

Mandell, J. F. 1975. Fatigue Crack Propagation Rates in Woven and Nonwoven Fiber Glass Laminates. Pp. 515-527 in Composite Reliability. Philadelphia: American Society for Testing and Materials. ASTM-STP 580.

Mandell, J. F. 1982. Fatigue Behavior of Fibre-Resin Composites. Pp. 67-107 in Developments in Reinforced Plastics, Volume 2, G. Pritchard, ed. London: Applied Science Publishers.

Mandell, J. F. 1990. Fatigue Behavior of Composite Materials. Chapter 7 in Fatigue Behavior of Short Fiber Composite Materials, K. L. Reifsnider, ed. New York: Elsevier.

Mandell, J. F., and U. Meier. 1983. Effects of Stress Ratio Frequency and Loading Time on the Tensile Fatigue of Glass-Reinforced Epoxy. Pp. 55-77 in Long Term Behavior of Composites, T. K. O'Brien, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 813.

Mandell, J. F., and J. Y. Tsai. 1990. Effects of Porosity on Delamination of Resin-Matrix Composites. Wright Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-89-3032.

Mandell, J. F., D. D. Huang, and F. J. McGarry. 1981. Tensile Fatigue Performance of Glass Fiber Dominated Composites. Composites Technology Review. 3(3):96-102.

Mandell, J. F., F. J. McGarry, A. J. Y. Hsieh, and C. G. Li. 1985. Tensile Fatigue of Glass-Fibers and Composites with Conventional and Surface Compressed Fibers. Polymer Composites. 6(3):168-174.

Martin, R. H., and G. B. Murri. 1988. Characterization of Mode I and Mode II Delamination Growth and Thresholds in Graphite/Peek Composites. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum 100577(April).

Miller, H. R., K. L. Reifsnider, W. W. Stinchcomb, D. A. Ulman, and R. D. Bruner. 1984. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-84-4007.

Miller, H. R., K. L. Reifsnider, W. W. Stinchcomb, D. A. Ulman, and R. D. Bruner. 1985. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-85-4069.

Miner, M. A. 1945. Cumulative Damage in Fatigue. Journal of Applied Mechanics. 12(3):A159-A164.

Nairn, J. A. 1989. The Strain Energy Release Rate of Composite Microcracking: A Variational Approach. Journal of Composite Materials. 23:1106-1129.

Narkis, M., E. J. H. Chen, and R. B. Pipes. 1988. Review of Methods for Characterization of Interfacial Fiber-Matrix Interactions. Journal of Polymer Composites. 9(4):245-251.

National Materials Advisory Board. 1968. Structural Design with Fibrous Composites. Washington, D.C.: National Materials Advisory Board. NMAB-236.

Noble, B., S. J. Harris, and M. J. Owen. 1983. Stress Corrosion Cracking of GRP Pultruded Rods in Acid Environments. Journal of Materials Science. 18(4):1244-1254.

O'Brien, T. K. 1986. Fatigue Delamination Behaviour of PEEK Thermoplastic Composite Laminates. Pp. 404-420 in Proceedings of the American Society for Composites: First Technical Conference, J. M. Whitney, ed. Lancaster, Pa.: Technomic.

O'Brien, T. K. 1988. Towards a Damage Tolerance Philosophy for Composite Materials and Structures. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum 100548(March).

O'Brien, T. K., I. S. Raju, and D. P. Garber. 1985. Residual Thermal and Moisture Influences on the Strain Energy Release Rate Analysis of Edge Delamination. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum 86437(June).

O'Brien, T. K., M. Rigamonti, and C. Zanotti. 1988. Tension Fatigue Analysis and Life Prediction for Composite Laminates. Hampton, Va.: National Aeronautics and Space Administration. Technical Memorandum 100549(October).

O'Brien, T. K., G. B. Murri, and S. B. Salpekar. 1989. Interlaminar Shear Fracture Toughness and Fatigue Thresholds for Composite Materials. Pp. 222-250 in Composite Materials: Fatigue and Fracture, P. A. Lagrace, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 1012.

Odom, E. M., and D. F. Adams. 1983. A Study of Polymer Matrix Fatigue Properties. Warminster, Pa.: Naval Air Development Center. Report No. NADC-83053-60.

Ogin, S. L., P. A. Smith, and P. W. R. Beaumont. 1985. Matrix Cracking and Stiffness Reduction During the Fatigue of a [0/90]s GFRP Laminate. Composite Science and Technology. 22:23-31.

Owen, M. J., and P. T. Bishop. 1973. Fatigue Properties of Glass-Reinforced Plastics Containing a Stress Concentrator. Journal of Applied Physics. 6(17):2057-2069.

Pagano, N. J., and R. B. Pipes. 1971. The Influence of Stacking Sequence on Laminate Strength. Journal of Composite Materials. 5:50-57.

Parvizi, A., K. W. Garrett, and J. E. Bailey. 1978. Constrained cracking in Glass Fibre-Reinforced Epoxy Cross-Ply Laminates. Journal of Materials Science. 13:195-201.

Phoenix, S. L., and R. G. Sexsmith. 1972. Clamp Effects in Fiber Testing. Journal of Composite Materials. 6:322-337.

Pless, W. M., S. M. Freeman, and C. D. Bailey. 1982. Advanced Methods for Damage Analysis in Graphite-Epoxy Composites. SAMPE Quarterly. 14(1):40-47.

Poursartip, A., M. F. Ashby, and P. W. R. Beaumont. 1985a. The Fatigue Damage Mechanics of a Carbon Fibre Composite Laminate: Development of the Model. Composites Science and Technology. 25:193-218.

Poursartip, A., M. F. Ashby, and P. W. R. Beaumont. 1985b. The Fatigue Damage Mechanics of a Carbon Fibre Composite Laminate: Life Prediction. Composite Science and Technology. 25:283-299.

Reddy, E. S., A. S. D. Wang, and Y. Zhong. 1987. Simulation of Matrix Cracks in Composite Laminates Containing a Small Hole. Pp. 83-91 in Damage Mechanics in Composites, A. S. D. Wang and G. K. Haritos, eds. New York: American Society of Mechanical Engineers.

Reifsnider, K. L. 1982. The Mechanics of Fatigue in Composite Laminates. Pp. 131-144 in Composite Materials: Mechanics, Mechanical Properties and Fabrication, K. Kawata and T. Akasada, eds. Englewood, N.J.: Applied Science Publishers.

Reifsnider, K. L. 1986. The Critical Element Model: A Modeling Philosophy. Journal of Engineering Fracture Mechanics. 25:739-749.

Reifsnider, K. L., and C. E. Bakis. 1986. Modeling Damage Growth in Notched Composite Laminates. Pp. 239-247 in Composites '86: Recent Advances in Japan and the United States, K. Kawata, S. Umekawa, and A. Kobayashi, eds. Tokyo: Japan Society for Composite Materials.

Reifsnider, K. L., and A. L. Highsmith. 1981a. Characteristic Damage States: A New Approach to Representing Fatigue Damage in Composite Materials. Pp. 246-260 in Materials Experimentation and Design in Fatigue. Guildford, U.K.: Westbury House.

Reifsnider, K. L., and A. L. Highsmith. 1981b. The Relationship of Stiffness Changes in Composite Laminates to Fracture Related Damage Mechanisms. Pp. 9-12 in Proceedings of the Second USA-USSR Symposium on Fracture of Composite Materials. Bethlehem, Pa.: Lehigh University Press.

Reifsnider, K. L., and W. W. Stinchcomb. 1986. A Critical-Element Model of the Residual Strength and Life of Fatigue-Loaded Composite Coupons. Pp. 298-313 in Composite Materials: Fatigue and Fracture, H. T. Hahn, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 907.

Rosen, B. W. 1964. Tensile Failure of Fibrous Composites. Journal of the American Institute of Aeronautics and Astronautics. 2(November):1985-1991.

Rotem, A. 1989. Stiffness Change of a Graphite Epoxy Laminate Under Reverse Fatigue Loading. Journal of Composites Technology and Research. 11(2):59-64.

Rotem, A., and H. G. Nelson. In press. Residual Strength of Composite Laminates Subjected to Tension-Compression Fatigue Loading. Journal of Composites Technology and Research.

Roylance, M., and D. Roylance. 1981. Effect of Moisture on the Fatigue Resistance of an Aramid-Epoxy Composite. Pp. 784-788 in Organic Coatings and Plastics Chemistry, Volume 45. Washington, D.C.: American Chemical Society.

Rummel, W. D., T. L. Tedrow, and H. D. Brinkerhoff. 1980. Enhanced X-Ray Stereoscopic NDE of Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-80-3053(June).

Russell, A. J. 1982. Factors Affecting the Opening Mode Delamination of Graphite/Epoxy Laminates. Defence Research Establishment Pacific. Materials Reports 82-Q.

Russell, A. J., and K. N. Street. 1987. The Effect of Matrix Toughness on Delamination: Static and Fatigue Fracture Under Mode II Shear Loading of Graphite Fiber Composites. Pp. 275-294 in Toughened Composites, N. J. Johnson, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 937.

Ryder, J. T., and K. N. Lauraitis. 1981. Effects of Load History on Fatigue Life. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-81-4155(December).

Ryder, J. T., and E. K. Walker. 1979. The Effect of Compressive Loading on the Fatigue Lifetime of Graphite/Epoxy Laminates. Air Force Materials Laboratory Technical Report. AFML-TR-79-4128(October).

Schultz, W. J., G. B. Portelli, R. C. Jordan, and W. L. Thompson. 1988. Fluorene Resins--A New Family of High Temperature Thermosetting Resins. Polymer Reprints. 29(1):136-137.

Schutz, D., and J. J. Gerhatz. 1977. Fatigue Strength of a Fiber Reinforced Material. Journal of Composites. 8(4):245-250.

Sendeckyj, G. P. 1981. Fitting Models to Composite Materials Fatigue Data. Pp. 245-260 in Test Methods and Design Allowables for Fibrous Composites, C. C. Chamis, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 734.

Sendeckyj, G. P. 1983. NDE Techniques for Composite Laminates. Pp. 2.1-2.22 in Characterization, Analysis, and Significance of Defects in Composite Materials, AGARD Conference Proceedings No. 355. Neuilly-sur-Seine, France: North Atlantic Treaty Organization Advisory Group for Aerospace Research and Development. AGARD-CP 355.

Sendeckyj, G. P. In press. Life Prediction for Resin-Matrix Composite Materials: In Fatigue of Composite Materials, K. L. Reifsnider, ed. New York: Elsevier Press.

Sendeckyj, G. P., G. E. Maddux, and E. Porter. 1982. Damage Documentation in Composites by Stereo Radiography. Pp. 16-26 in Damage in Composite Materials. K. L. Reifsnider, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 775.

Stinchcomb, W. W., and K. L. Reifsnider. 1986. Cumulative Damage Model for Advanced Composite Laminates. Center for Composite Materials and Structures, Virginia Polytechnic Institute and State University Technical Report. Report CCMS-86-05(August).

Talreja, R. 1985. Fatigue of Composite Materials. Ph.D. dissertation. Technical University of Denmark.

Tamuzs, V. 1982. Some Peculiarities of Fracture in Heterogeneous Materials. Pp. 131-137 in Fracture of Composite Materials, G. C. Sih and B. P. Tamuzs, eds. The Hague: Martinus Nijhoff Publishers.

Tirosh, J. 1973. The Effect of Plasticity and Crack Blunting on the Stress Distribution in Orthotropic Composite Materials. Journal of Applied Mechanics. 40:785-790.

Trethwey, B. R., J. W. Gillespie, and L. A. Carlsson. 1988. Mode II Cyclic Delamination Growth. Journal of Composite Materials. 22:459-483.

Waddoups, M. E. 1968. Characterization and Design of Composite Materials. Pp. 254-308 in Composite Materials Workshop, S. W. Tsai, J. C. Halpin, and N. J. Pagano, eds. Westport, Conn.: Technomic Publishing.

Waddoups, M. E., J. R. Eisenmann, and B. E. Kaminski. 1971. Macroscopic Fracture Mechanics of Advanced Composite Materials. Journal of Composite Materials. 5:446-454.

Wang, A. S. D. 1984. Fracture Mechanics of Sublaminate Cracks in Composite Laminates. Composite Technology Review. 6:45-62.

Wang, A. S. D., and F. W. Crossman. 1980. Initiation and Growth of Transverse Cracks and Edge Delamination in Composite Laminates, Part I, An Energy Method. Journal of Composite Materials. 4:71-87.

Wang, A. S. D., P. C. Chou, C. S. Lei, and R. B. Bucinell. 1984. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-84-4004(Phase II).

Wang, A. S. D., P. C. Chou, C. S. Lei, and R. B. Bucinell. 1985. Cumulative Damage Model for Advanced Composite Materials. Wright-Patterson Air Force Base, Ohio: Air Force Wright Aeronautical Laboratories. AFWAL-TR-85-4104.

Wang, S. S. 1981. Edge Delamination in Angle-Ply Composite Laminates. Pp. 473-484 in Proceedings of the 22nd Structures, Structural Dynamics, and Materials Conference, Part I. New York: American Institute of Aeronautics and Astronautics.

Weibull, W. 1939. A Statistical Theory of the Strength of Materials. Proceedings of the Royal Swedish Institute of Engineering Research. 151:1-45.

Whitney, J. M. 1983. Residual Strength Degradation Model for Competing Failure Modes. Pp. 225-245 in Long Term Behavior of Composites, T. K. O'Brien, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 813.

Wilkins, D. J. 1981. A Comparison of the Delamination and Environmental Resistance of a Graphite-Epoxy and a Graphite-Bismaleimide. Naval Air Systems Command Technical Reports. NAV-GD-0037.

Wilkins, D. J. 1983a. Damage Tolerance of Composites: Changing Emphasis in Design, Analysis, and Testing. In Proceedings of the Sixth Conference on Fibrous Composites in Structural Design.

Wilkins, D. J. 1983b. The Engineering Significance of Defects in Composite Structures. Pp. 20-21 in Characterization, Analysis, and Significance of Defects in Composite Materials, AGARD Conference Proceedings No. 355. Neuilly-sur-seine, France: North Atlantic Treaty Organization for Aerospace Research and Development. AGARD-CP 355.

Wilkins, D. J., J. R. Eisenmann, R. A. Camin, W. S. Margolis, and R. A. Benson. 1982. Characterizing Delamination Growth in Graphite-Epoxy. Pp. 168-183 in Damage in Composite Materials, K. L. Reifsnider, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 775.

Wolff, R. V., and G. H. Lemon. 1972. Reliability Prediction for Adhesive Bonds. Air Force Materials Laboratory Technical Report. AFML-TR-72-121.

Wolff, R. V., and D. J. Wilkins. 1980. Durability Evaluation of Highly Stressed Wing Box Structure. Pp. 761-770 in Fourth Conference on Fibrous Composites in Structural Design, E. M. Lenoe, D. W. Oplinger, and J. J. Burke, eds. New York: Plenum Press.

Yang, J. N. 1977. Reliability Prediction for Composites Under Design Proof Tests in Service. Pp. 272-295 in Composite Materials: Testing and Design (Fourth Conference). Philadelphia: American Society for Testing and Materials. ASTM-STP 617.

Yang, J. N. 1978. Fatigue and Residual Strength Degradation for Graphite/Epoxy Composites Under Tension-Compression Cyclic Loadings. Journal of Composite Materials. 12:19-39.

Yang, J. N., and R. T. Cole. 1982. Statistical Analysis of Composite Fatigue Life. Pp. 333-340 in Science and Engineering of Composites, T. Hayashi et al., eds. Tokyo: Japan Society for Composite Materials.

Yang, J. N., and D. L. Jones. 1978. Statistical Fatigue of Graphite/Epoxy Angle Ply Laminates. Journal of Composite Materials. 12:371-389.

Yang, J. N., and D. L. Jones. 1980a. Statistical Fatigue of Unnotched Composite Laminates. Pp. 472-483 in Advances in Composite Materials (ICCM3), Volume 1, A. R. Bunsell et al., eds. London: Pergamon Press.

Yang, J. N., and D. L. Jones. 1980b. The Effect of Load Sequence on Statistical Fatigue of Composites. AIAA Journal. 18(12):1525-1531.

Yang, J. N., and D. L. Jones. 1981. Load Sequence Effects on the Fatigue of Unnotched Composite Materials. Pp. 213-232 in Fatigue of Fibrous Composite Materials, K. N. Lauraitis, ed. Philadelphia: American Society for Testing and Materials. ASTM-STP 723.

Yang, J. N., and D. L. Jones. 1982. Fatigue of Graphite/Epoxy [0/90/+45]s Laminates Under Dual Stress Levels. Composites Technology Review. 4(3):63-70.

Yang, J. N., and M. D. Liu. 1977. Residual Strength Degradation Model and Theory of Periodic Proof Tests for Graphite/Epoxy Laminates. Journal of Composite Materials. 11:176-203.

Yang, J. N., and C. T. Sun. 1980. Proof Test and Fatigue of Unnotched Composite Laminates. Journal of Composite Materials. 14:168-176.

Yang, J. N., R. K., Miller, and C. T. Sun. 1980. Effect of High Load on Statistical Fatigue of Unnotched Graphite/Epoxy Laminates. Journal of Composite Materials. 14:82-94.

Yang, J. N., S. H. Yang, and D. L. Jones. In press. A Stiffness-Based Statistical Model for Predicting the Fatigue Life of Graphite/Epoxy Laminates. Journal of Composites Technology and Research.

Yee, A. F. 1984. Toughening Mechanisms in Elastomer-Modified Epoxy Resins, Part 2. Hampton, Va.: National Aeronautics and Space Administration. NASA Technical Memorandum 3852(December).

Appendix

BIOGRAPHICAL SKETCHES OF COMMITTEE MEMBERS

MAX E. WADDOUPS is president of Marine Systems Supply and has accumulated nearly 30 years of composite mechanics experience. Prior to leaving General Dynamics, he was program director for the Air Force AFTI/F-16 technology demonstrator project and the National Aerospace Plane program. He received his B.S. degree from Westminster College and his M.S. degree in mechanical engineering from Brigham Young University.

T. T. CHIAO received his B.S. degree from the Taiwan College of Engineering and his M.S. degree in chemical engineering from the University of Rhode Island. He is retired from the Lawrence Livermore National Laboratory and is now a consultant. He is also a member of the Educational Board of the Chinese Journal of Composite Materials and the Advisory Board of the Journal of Composite Materials.

MELVIN F. KANNINEN received his B.S. degree from the University of Minnesota and his M.S. and Ph.D. degrees from Stanford University. Dr. Kanninen is currently Program Director of Engineering Mechanics at the Southwest Research Institute. He is a member of the National Academy of Engineering and the National Materials Advisory Board.

JOHN F. MANDELL is a professor at Montana State University in the chemical and mechanical engineering departments. Professor Mandell received his S.B. degree from the University of Massachusetts at Amherst, his S.M. from Case Western Reserve University, and his Ph.D. from the Massachusetts Institute of Technology.

KENNETH L. REIFSNIDER is professor of engineering mechanics and chairman of the Materials Engineering Science Program at Virginia Polytechnic Institute and State University. He received his B.A. from Western Maryland College and his B.E.S. and Ph.D. (in metallurgy) from Johns Hopkins University.

CHARLES W. ROGERS is a research structures engineer at Bell Helicopter, Textron, Inc. He received his B.S. degree in mechanical engineering from Texas A&M University and his M.S.E. from Southern Methodist University. He has been active in the field of composite materials since 1960. He has been a member of the National Aeronautics and Space Administration's Structures and Materials Advisory Group, the editorial staff of the Journal of Composite Materials, and the Ratheon Starship Review Committee. He has several patents in the composites field.

JAMES T. RYDER is manager of the Materials Quality and Processing Department, Materials Sciences Directorate, Research and Development Division, Lockheed Missiles and Space Company. Dr. Ryder received his B.S., M.S., and Ph.D. (theoretical and applied mechanics) from the University of Illinois. He is a member of Sigma Xi.

SANFORD S. STERNSTEIN is William Weightman Walker Professor of Polymer Engineering and director of the Center for Composite Materials and Structures at Rensselaer Polytechnic Institute. He received his B.S. from the University of Maryland and his Ph.D. in chemical engineering from Rensselaer Polytechnic Institute. He is a fellow of the American Physical Society.

DICK J. WILKINS is president of the Delaware Technology Park and professor of mechanical engineering at the University of Delaware. He received his B.S., M.S., and Ph.D. degrees from the University of Oklahoma. He also served as president of the American Society for Composites in 1990-1991.